

Metal Progress

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Cover Credit: Victor Keppler photographed the handsome jewelry and tableware made of stainless steel in a Christmas setting. Engravings were loaned by Electro Metallurgical Co., unit of Union Carbide and Carbon Corp.

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The Austenitic Stainless Steels— American and British Practice Compared

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In comparing American and British practice concerning 18% chromium, 8% nickel stainless steels, the author's purpose is "to invite discussion that may give positive guidance to those in both countries who direct the manufacture and use of these important and costly materials". The principal points of difference in British and American practice concern the use of unstabilized compositions, the safe upper limit of carbon content in stabilized steels, and the use of the Huey test for corrosion.

THE FUNDAMENTAL IMPORTANCE of the austenitic stainless steels to many industries justifies a great deal of research and investigation on the problems of the right utilization of existing knowledge and the constant extension of knowledge. All the major industrial countries have contributed to the pool of knowledge on which present practice is based, and it may be instructive and informative to examine the use made of existing knowledge by two of the major contributors—the United States and Britain. This comparison is based on some years of experience as a user, on discussions with American producers and users, and on some familiarity with the relevant technical publications of the two countries. Its sole purpose is to invite discussion that may give positive guidance to those in both countries who direct the manufacture and use of these important and costly materials.

Present Position—A direct comparison between the outputs of chromium-nickel austenitic steels in

the two countries is difficult. Published statistics are compiled on different bases; without access to the sources of both sets of figures, accurate coordination of the statistics is not possible. For the purpose of this discussion, however, it is not imperative to have accurate figures; the relevant fact is the much greater production in the United States. If the British figure is taken as 30,000 tons per year, the American figure will be of the order of five or six times greater.

Since the economics of the producing industries are not under discussion here, it is sufficient to say that the large American output, reflecting the demands of larger industries, also indicates a more widespread utilization of these special materials. It is probable that the total U.S. figure contains an adequate outlet for the scrap of that part of the production made to close specification requirements,

the outlet consisting of the extensive use of these materials for decorative and domestic use, in which technical requirements are not so stringent. In Britain, the lower production and less widespread demand result in a less balanced production and this state of affairs may explain certain important differences in composition of the high-grade products of the two industries.

In both countries, a significant proportion of output is made to close specification requirements and goes to users whose experience demands close control over composition and treatment. These users, generally large producing industries, are largely instrumental in fixing the limits of national or other specifications to which the steelmakers must work. It is with this particular part of the total production that this discussion is concerned and with that portion of it which includes and is developed from the basic composition 18% chromium, 8% nickel.

In studying the production figures for both countries, one significant fact emerges immediately — that is, the much smaller use that is made of stabilized compositions (compositions immune to intercrystalline disintegration) in the United States as compared with Britain. It would appear that about 5% of the total production is in this category, as compared with a very much higher proportion in Britain. There is thus a strong presumption that large quantities of unstabilized material are still being used in the United States. Some of this material must be used in the welded condition and, therefore, must be subjected to heat treatment before use. Here is the first significant point of difference between the two countries, since the practice of heat treating such welded structures has virtually disappeared in Britain.

A second difference in practice between the two countries can be noted from current specifications. In America, the tendency is definitely to lower carbon contents and this tendency is now being followed in the production of a steel of total carbon less than 0.03%.

A third difference is noted in the group of special steels containing molybdenum. These valuable steels have been in use in both countries for many years. At the present time, a significant proportion of the British production is stabilized; little or none of the American production contains a stabilizing addition and again resort is had to final heat treatment.

A final difference to be noted lies in the choice of stabilizing element. In America, columbium is in strong preference; in Britain, the bulk of the stabilized material contains titanium.

This brief summary indicates the major differences in practice between the two countries. It is proposed now to examine these differences a little more closely.

Unstabilized Compositions

It will be a matter of general agreement that, if at all possible, post-welding heat treatment should be avoided. At best, in small and simple structures and fittings, it is a costly and troublesome business, with its concomitant descaling. At worst, in large complicated structures, it is a major operation, demanding special equipment and close control. To heat a large, unsymmetrical fitting to a temperature in the region of 1100° C. (2000° F.) and to provide for cooling fast enough to prevent carbide separation is an operation not to be undertaken lightly. When the added complication is considered — residual stress caused by unequal cooling rates, frequently developing into actual distortion — the vigorous prosecution

of the research that produced the stabilizing elements seems more than justified. This research has produced such elements, and their efficiency and value have been amply proved by years of experience in both countries. In Britain, this experience has been interpreted as justifying the complete cessation of the post-welding quenching treatment and the almost exclusive use of stabilized compositions for industrial equipment. In the United States, this development has not been so generally adopted, and in view of the difficulties and uncertainties of the alternative procedure, one naturally looks for the basis of choice of this alternative.

The principal justification — and perhaps the only one — appears to be the results of comparative corrosion losses in the test that seems to be a widely accepted standard of measurement in the United States — the Huey test. Many published results of this test show that the water quenched, unstabilized material has lower rates of corrosion loss than the alternative stabilized material. This observation is merely noted at this stage, and will be referred to later. A real cost comparison of the alternatives is difficult. A direct comparison of the monetary cost of the same piece of equipment in the two alternative constructions — stabilized as against unstabilized and heat treated — may show a lower figure for the latter. One might be puzzled, however, as to the addition one should make to the actual cost in money of the heat treated component to cover the risks involved in its use. These risks consist of using an internally stressed component or a component in which the cooling rate in quenching had not been quite so high as that employed in the tests on which its use was based. On the matter of residual stress it is impossible, of course, to follow the quenching with stress relieving, for stress relieving would necessarily be done at some temperature that would produce carbide precipitation.

Carbon Contents

A comparison of the specifications in current use in the two countries will indicate at once that higher limits of carbon content are tolerated in Britain. Such higher limits (up to about 0.16%) have been in general use in many industries for a number of years. British steelmakers would view with some concern any move to force these limits to a lower level and would point to the many years of satisfactory service given by the steels of higher carbon content. In the United States, specification limits are set in the region of 0.06 to 0.08%, and it is possible that these limits meet with no opposition from the steelmakers. Further, these are no

doubt the carbon limits on which a mass of corrosion data has been accumulated. It may also be recorded that, in most steels of the basic commercial compositions, higher corrosion losses are shown in the Huey test by steels having carbon at the British, rather than the American, levels.

Steels Containing Molybdenum

For many years American users of austenitic steels containing molybdenum were supplied with a steel containing no stabilizing element. This observation would appear to be true of current practice, although there are indications of a change in this direction. It is important to note that this practice was developed and was in general use for some years before research had shown the dangers of the sigma phase in this particular composition. Before the dangers of the sigma phase were realized, American practice had developed in the direction of utilizing the unstabilized steels containing molybdenum and heat treating the finished structures as was done with the simpler, unstabilized steels. Sometimes the dangers and difficulties of the heat treatment were recognized by placing limits on the thickness of material that should be heat treated. It seems fairly certain that this earlier practice was based on the demonstrable superiority of the unstabilized material in the Huey test.

It is, of course, now well known that the addition of a stabilizing element—either columbium or titanium—to an austenitic stainless steel containing molybdenum strongly increases the tendency to sigma formation. If, however, real economic advantage derives from the use of the stabilized material, this adventitious discovery should not be allowed to perpetuate the older practice and should not divert attention from the possibilities of evolving a composition containing adequate molybdenum for satisfactory resistance in most recognized environments in which this steel is used, a composition effectively stabilized and immune to sigma formation in ordinary commercial production and fabrication processes. Much progress in evolving such a steel has already been made, a good deal of it in the United States.

Titanium or Columbium for Stabilization?

The first element to be successfully and extensively used for stabilization of this class of steel was titanium, and many thousands of tons of titanium-stabilized 18% chromium, 8% nickel steel were used all over the world. Early application of the Huey test to this steel showed that its corrosion losses were higher than those of the plain,

unstabilized material. These results may well have put a brake on expanding use in the United States. In Britain, where the Huey test was not used as a general performance test, no arrest in development occurred, and titanium came into wide, general use. Some attempt was made in the United States to improve the Huey test performance by applying the "stabilizing" heat treatment, but, of course, the introduction of a heat treatment into the production line was a definite handicap to development.

In these circumstances, columbium was introduced as an alternative stabilizing element. Generally, it was used with a somewhat higher nickel content than was normal for the titanium-stabilized steel. The higher content of nickel would counteract the strong ferrite-forming characteristic of the columbium, and a steel approaching the single-phase structure resulted. Such a steel gives, no doubt, a better performance in the Huey test than the markedly duplex titanium-stabilized steel. Comparison of the two varieties of stabilized steel in a wide range of conditions of corrosion would be a valuable contribution to knowledge. Such results as are available to the author indicate general similarity in performance, with one established condition in which the columbium-stabilized material is superior. Further points to be noted are the superiority in creep strength shown by the columbium-stabilized steel in certain conditions of temperature and stress and its more satisfactory behavior in polishing.

In Britain, the titanium-stabilized steel is still widely used and will probably continue to be so used, because of its excellent performance for twenty years in many industrial applications. At the same time, it is recognized that the newer columbium-stabilized steel has superiority in a few of the conditions in which the two have been compared. There is thus, at the present state of knowledge, need for both types. It may be well to ask, however, whether the long-term view would not be to encourage the use of the element that has given good performance over a period of years and that is in abundant supply in the earth's crust, rather than a material that may become scarce.

The Huey Test

It may be appropriate, at this stage, to set down in detail the standards of assessment and comparison used in the two countries.

In the United States, the Huey test is in apparent general use. This test consists of subjecting the steel under examination to contact with boiling 65% nitric acid for one or more periods of 48 hr., under carefully standardized and controlled conditions. Performance is assessed by loss in weight.

Increasing use is made of another test, described below, both in investigations and in specifications.

In Britain, the test in general use is the copper sulphate-sulphuric acid test first used by W. H. Hatfield, but often referred to as the Strauss test. The test is used in two ways. If a new and untried composition is under examination, the samples are subjected for a period of 50 days to the boiling $\text{CuSO}_4 - \text{H}_2\text{SO}_4$ mixture, the conditions being carefully standardized and controlled. If the material under test has a composition within the limits of a standard specification—that is, if the composition is a proven one—the test is applied for only 72 hr. In both uses of the test, assessment is by bending and microscopic examination. These quality tests are always supplemented by direct corrosion tests for specific applications, the criterion in such tests being, of course, loss in weight.

In comparing the two testing procedures, Huey and Strauss, it is important to note that, in both, the characteristic of the steel under examination is susceptibility to intergranular attack. The actual corrosion performance, in the generally accepted sense, is not being measured. It must, therefore, be assumed in the Huey test, that no corrosion will occur except at susceptible grain boundaries. The use of a powerful oxidizing material as the test medium is some guarantee that this assumption is justified and that, further, the rate of corrosion at active anodic areas will be accentuated by suppression of any potential local anodes. Nevertheless, if any corrosion does occur at regions remote from grain boundaries (for example, at the site of nonmetallic impurities) the loss will be attributed to grain-boundary attack and the steel under examination will be falsely assessed on this basis. The well-known erratic behavior of these steels in any corrosion test raises some doubts as to whether the assumption of the Huey test is justified in every test undertaken. When one has seen, in innumerable corrosion tests, duplicate samples of many compositions behave in totally different manner

when tested in the same solution, one is very chary of making any assumption about the corrosion behavior of these steels. The copper sulphate-sulphuric acid test is not beyond criticism. It is confined, however, to examining the performance of grain boundaries under searching conditions and it attempts to do so by direct tests that are confined to assessment of grain-boundary conditions, without assumption.

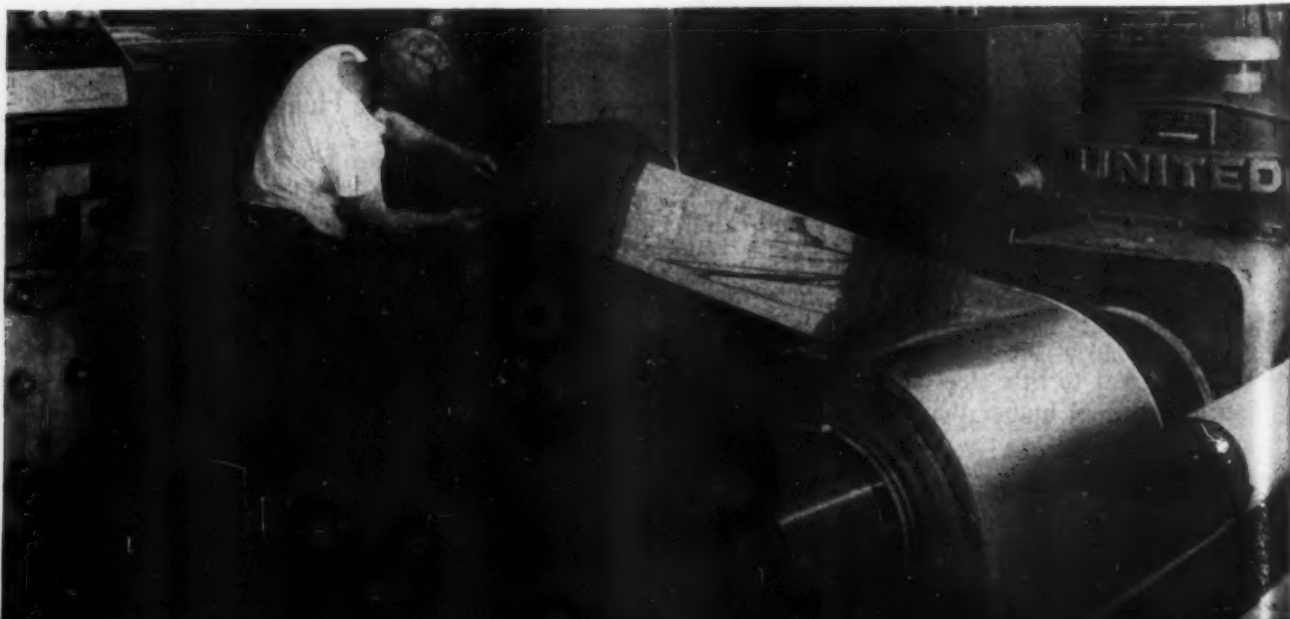
Queries

The investigator, striving to be unbiased and to get the right answers to questions involved in the use of these steels, must pose these queries:

1. What is the justification for the continued use of unstabilized steels, with their concomitant heat treatment?
2. In what circumstances is it permissible to use unstabilized steel containing molybdenum?
3. What is the safe upper limit of carbon content in stabilized steels?
4. Is the assumption involved in the use of the Huey test valid? Has the test been pushed too far as a means of assessment of fitness for general industrial use?

Conclusion

The author of this article is, naturally, more familiar with conditions in his own country than those in the United States. For any unjustifiable assumptions regarding American practice, he would wish to apologize and plead the errors of the inquiring stranger. To American friends with whom he has discussed some of the difficulties in reconciling practice in the two countries it is unnecessary to state that he would not presume to criticize and is actuated only by a desire for the truth; to others he would make this assurance, in the hope that a frank exchange of views is the best way of resolving differences. ❶



Retained Austenite

Abstract of Campbell Memorial Lecture

By Morris Cohen

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The twenty-third Campbell Memorial Lecture was a comprehensive analysis of the theory and data concerning retained austenite in steel. This abstract emphasizes the main facts about retained austenite, but omits the discussion of theory. The complete document will appear in Volume 41 of the Transactions, to be published early in 1949.

IT IS a familiar fact that austenite decomposes into ferrite, carbide, pearlite, bainite, or martensite, depending on the conditions of cooling and the composition of the steel. Because most of the austenite is consumed in the transformation, we may often regard it solely as a means to an end. However, in the hardening of steel, part of the austenite frequently survives the quenching that leads to martensite formation, and remains with the martensite at room temperature. Such retained austenite may have important effects on the properties of the steel, both during tempering and in later service.

To treat the problem of retained austenite in a fundamental way, we must consider the following questions: How does retained austenite originate? What are the factors governing its amount? Why is it so persistent? What happens to it during aging at room temperature, during cooling below room temperature, and during tempering above room temperature? How does it respond to stress and strain?*

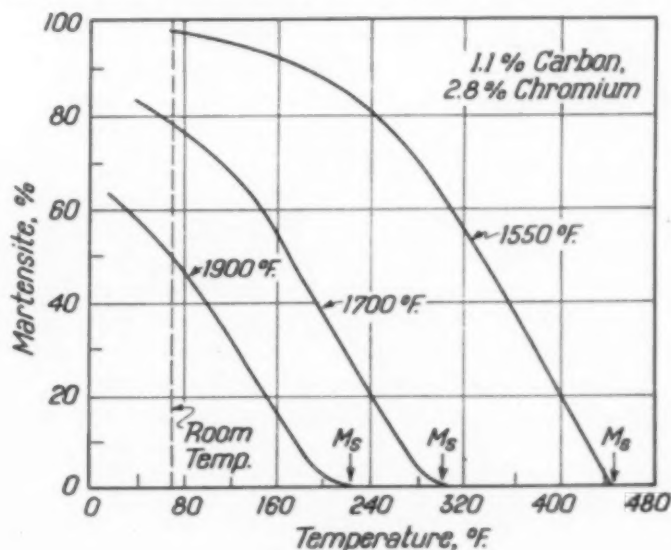
In the hardening of steel, the desired transformation product is martensite. Therefore, the steel is cooled fast enough to prevent the austenite from decomposing into any of the other

products, and it is brought to the low-temperature range where martensite forms.

Unlike other transformations in steel, the martensite reaction cannot be suppressed by rapid cooling, and the reaction proceeds only while the temperature is being lowered below M_s . The quantitative course of the austenite-martensite transformation during cooling below M_s is shown in Fig. 1, for a steel containing 1.1% carbon and 2.8% chromium, austenitized at three different temperatures. The intercepts on the dashed vertical line indicate the amounts of martensite and (by subtraction)

retained austenite at room temperature after these austenitizing and cooling treatments. Room temperature is merely a convenient stopping point for the hardening quench in practice. Actually, the transformation curves extend smoothly to

Fig. 1 — Martensite Transformation Curves for Continuous Cooling of a Steel Containing 1.1% Carbon, 2.8% Chromium. Specimens were austenitized at 1550, 1700 and 1900° F., as indicated. (W. J. Harris and M. Cohen)



*EDITOR'S NOTE: In his lecture, Professor Cohen did consider all these questions. However, the first and last items are largely omitted here.

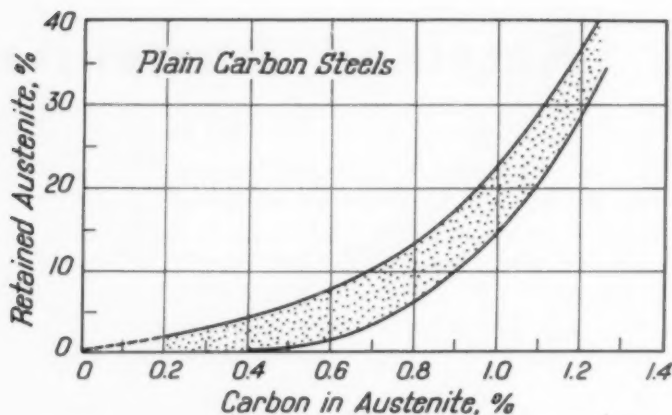


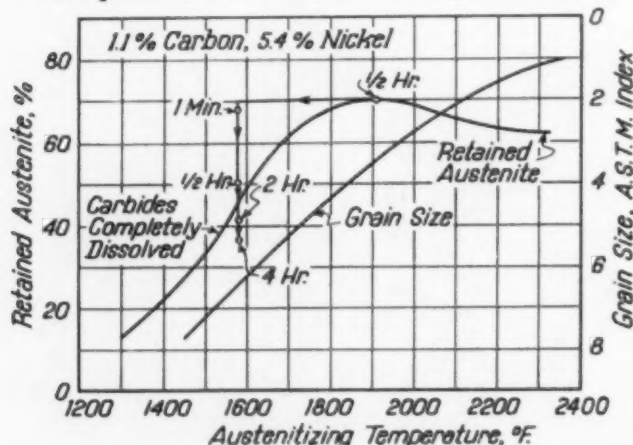
Fig. 2 — Effect of Carbon Content on Retained Austenite in Plain Carbon Steel. All specimens were austenitized above A_{c3} or A_{cm} and quenched in water or brine.

subzero temperatures. Subzero transformation is essentially self-stopping; that is, it peters out short of completion — even during continuous and rapid cooling. Plain carbon and low-alloy steels (0.70 to 1.25% carbon) stop transforming at about -250°F . with from 0.5 to 5% austenite still available. High speed steel and high-carbon high-chromium steel stop transforming at about -150°F . with from 8 to 15% austenite remaining.

Quantities of Retained Austenite

Retained austenite contents at room temperature range all the way from 100% in the austenitic steels down to virtually zero in the plain low-carbon steels. Stainless steel of the 18-8 type and Hadfield's steel containing 1.2% carbon, 12.5% manganese are the most familiar examples of

Fig. 3 — Effect of Austenitizing Temperature on Retained Austenite in a Nickel Steel Containing 1.1% Carbon, 5.4% Nickel. All specimens were austenitized for $\frac{1}{2}$ hr., except as indicated otherwise, and quenched in brine at 70°F . (S. A. Kulin)



commercial austenites having M_s below room temperature. We find from 25 to 35% retained austenite in commercially hardened 1% carbon, 5% chromium die steel, 15 to 30% austenite in the 1.5% carbon, 12% chromium die steels, 15 to 25% austenite in high speed steels, and 3 to 15% in low-alloy toolsteels. Commercially hardened, plain carbon toolsteels retain about 5 to 10% austenite.

As shown in both Fig. 1 and 2, the quantity of retained austenite depends greatly on the dissolved carbon content. It is surprising that even in 0.2% carbon steel (brine quenched from above A_3), we find up to 2% retained austenite.

Effect of Austenitizing Temperature. It is well known that the austenitizing temperature has a large effect on M_s , and therefore on austenite retention, because the composition of the austenite changes with increasing carbide (or ferrite) solution. This factor is of major importance in the commercial heat treatment of high-carbon steels, where some of the carbide is deliberately left undissolved.

There is another effect of austenitizing temperature distinct from that of composition change. This phenomenon can be observed particularly well in the 5% nickel steel that we shall use as an example. In Fig. 3, we note that the retained austenite content continues to increase by a large amount even when the austenitizing temperature is higher than is necessary to dissolve all the carbon. Austenitizing at 1550°F . for $\frac{1}{2}$ hr. dissolves all the carbon in this 1.1% carbon, 5.4% nickel steel, and 39% austenite is retained after brine quenching. Yet after austenitizing at 1900°F . and quenching, there is 70% retained austenite without any further change in the composition of the austenite. This phenomenon cannot be explained by assuming that submicroscopic carbides have dissolved or that concentration gradients have been leveled, because the effect is essentially reversible. If the steel is first austenitized at 1900°F ., then cooled to 1580°F . (above the carbide precipitation temperature) and held long enough, the percentage of retained austenite gradually decreases and overtakes the value that is obtained when the steel is austenitized at 1580°F . in the first place.

The decrease in retained austenite after austenitizing above 1900°F . appears to be the result of increasing grain size, which favors the formation of martensite. The effect is rather small compared with the other austenitizing variables and is obscured by them, except at the higher temperatures after the other factors have expended themselves somewhat. The grain size effect also accounts

for the fact that, after the steel has been held for a long time at 1580° F. following austenitizing at 1900° F., the retained austenite decreases *below* the value obtained by direct austenitizing at 1580° F.

Stabilization

Figure 4 shows the effect of interrupting the hardening quench of a 1.1% carbon, 1.5% chromium steel for 30 min. at various temperatures within the martensite range before further cooling. If, for example, the hardening quench is stopped at 130° F., the steel contains 51% martensite on reaching this temperature. After holding for 30

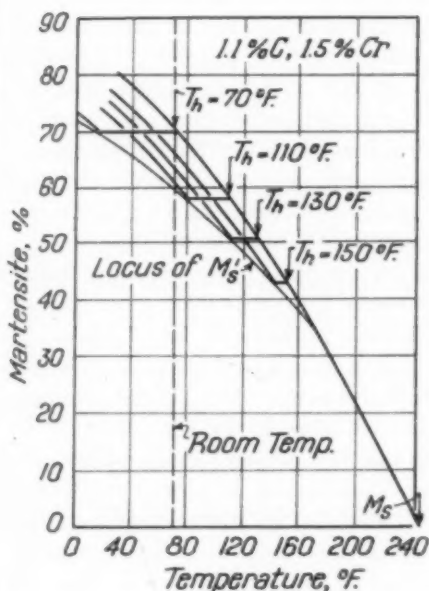


Fig. 4 — Martensite Transformation Curves for Continuous and Interrupted Cooling of a Chromium Steel Containing 1.1% Carbon, 1.5% Chromium. All specimens were austenitized at 1900° F. (W. J. Harris and M. Cohen)

min., the cooling transformation is completely inhibited down to about 115° F., where it sets in again and follows parallel to the normal transformation curve. During the cooling between 130 (T_h , the holding temperature) and 115° F. (M_s'), approximately 6% of the austenite would have transformed to martensite, were it not for the stabilizing effect at T_h . This 6% remains lost to the transformation; thus the curve for the stabilized austenite is displaced vertically from the curve for the unstabilized austenite by just this amount.

If hardened steel is aged at room temperature before being subcooled, the retained austenite becomes stabilized against subzero transformation.

The characteristics of this process are shown in Fig. 5. In the first place, some isothermal decomposition of the retained austenite occurs while the steel is at room temperature. This is shown by the short vertical line at 68° F. The course of the subsequent cooling transformation is shown after 1, 10 and 50 hr. at room temperature, in comparison with the normal transformation for continuous cooling through room temperature. It is evident that, as a result of the stabilizing process, there is a temperature interval for each treatment, down to the point labeled M_s' , during which no detectable transformation occurs. Below M_s' , the austenite-martensite reaction sets in again, and follows

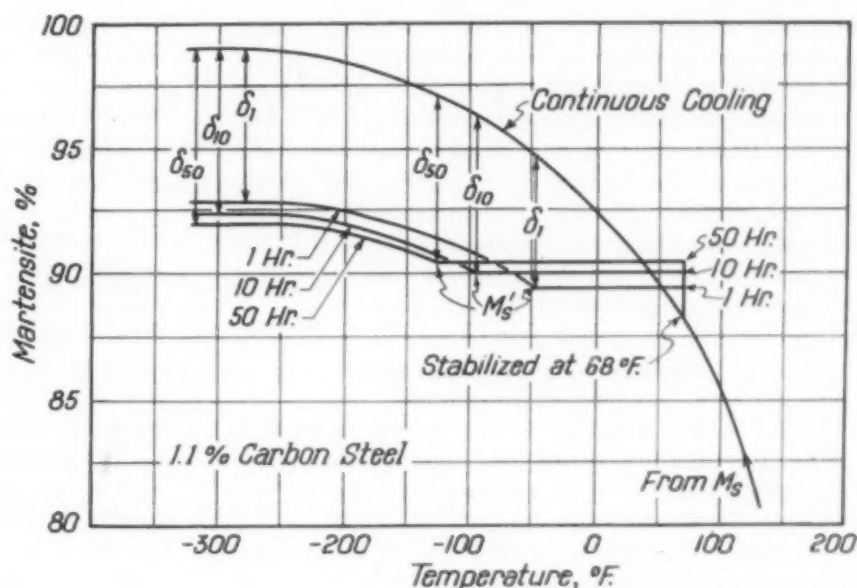


Fig. 5 — Effect of the Duration of Room-Temperature Aging on the Stabilization of Austenite, as Shown by Martensite Transformation Curves for Continuous and Interrupted Cooling of a 1.1% Carbon Steel. All specimens were austenitized at 1450° F. Cooling was interrupted by holding specimens for 1, 10 and 50 hr. at room temperature before cooling to subzero temperature. The δ -values indicate the extent of stabilization.

nearly parallel to the continuous cooling curve, but displaced from it by approximately the amount of transformation (δ) that would occur between room temperature and M_s' if the stabilization had not taken place. The curves for stabilized austenite flatten out at the same temperature as does the curve for unstabilized austenite, and this indicates that the stabilization does not shift the transformation curve to lower temperatures. Rather, there is an almost quantitative displacement in the vertical direction, just as though the percentage of austenite that does not transform on cooling to M_s' remains lost to the transformation, even when the steel is cooled to temperatures well below

M_s' . The quantity δ , therefore, is a measure of the extent of stabilization. It should be emphasized that the amount of isothermal transformation taking place at room temperature is much too small to account for the stabilizing phenomenon.

The quantity δ is larger, and M_s' is depressed more, the longer the holding time at room temperature. If the hardened steel is aged at elevated temperatures, instead of at room temperature, before refrigeration, the stabilization is even more marked and occurs more rapidly.

Because stabilization depends on both time and temperature, it may occur during retarded cooling. This explains the fact that oil quenching of steel frequently leads to more retained austenite than water quenching. Martempering, which involves air cooling through the martensite range, results in still more retained austenite.

When the austenite is highly alloyed, the tendency toward stabilization becomes so marked that it can occur even in the absence of martensite. For example, in certain high-carbon high-chromium steels, stabilization sets in if the cooling rate in approaching M_s is retarded ever so slightly; the steel must be cooled extremely fast to prevent M_s from being depressed by stabilization.

Bainite Formation Below M_s

When the cooling of steel is interrupted at some temperature below M_s , any austenite that has not transformed to martensite becomes available for isothermal decomposition to bainite.

Above M_s , the rate of bainite formation decreases with decreasing temperature in accordance with the well-known TTT-curve. Just below M_s , the bainite reaction is accelerated by the co-existing martensite, but this stimulating effect fades out at longer times and at lower temperatures. Near room temperature, the retained austenite is so sluggish with respect to bainite formation that bainite does not form in any detectable amount for thousands of hours. At these low temperatures, retained austenite is remarkably persistent.

Tempering

Aging at Room Temperature. Some of the retained austenite in hardened steel does decompose during aging at room temperature. We referred to this briefly in connection with Fig. 5, when it was mentioned that the amount of transformation at room temperature is much too small to account for the stabilizing phenomenon. Though slight, this transformation at room temperature is nevertheless extensive enough to cause dimensional

changes, and these dimensional changes are sometimes great enough to impair the accuracy of gages made from a steel that contains a few per cent of retained austenite after tempering.

The room-temperature decomposition of retained austenite starts as soon as the steel reaches the temperature of the quenching medium, and continues for many weeks. Usually, however, less than 5% of the austenite transforms, and about half of that during the first two hours after the quench. The rate of this reaction is quite insensitive to the aging temperature; it proceeds at about the same rate anywhere between 32 and 200° F.

Tempering at Elevated Temperatures. Above about 250° F., the product of retained austenite decomposition is exactly the same as the product that forms isothermally from primary austenite at the same temperature. However, the rate of decomposition of retained austenite is usually affected by the co-existing martensite. For instance, the time for the completion of the bainite reaction is shorter for retained austenite than for primary austenite in the same steel. The TTT-curve for transformation of retained austenite shows a bainite "nose" at the same temperature as for transformation of primary austenite, and there is similar correspondence in the carbide precipitation and the formation of pearlite from retained and primary austenites of similar composition.

Cooling From the Tempering Temperature. In tempering, we are concerned not only with the products of transformation at the tempering temperature, but also with reactions during cooling from the tempering temperature. Not all the retained austenite transforms during tempering, but something important may happen to the part of it that does not transform. For instance, when 18-4-1 high speed steel is tempered for 1 hr. at 600° F., there is no detectable transformation, but stabilization occurs and no martensite forms during subsequent cooling to -300° F. However, if this steel is tempered at 1050° F., M_s' is raised from room temperature to 470° F.; that is, the retained austenite can transform to martensite during cooling from the tempering temperature. Carbide precipitation, which decreases the carbon content of austenite at the tempering temperature, helps to raise M_s' but it is not the only factor in "conditioning" the austenite to transform. In the nickel steel of Fig. 3, conditioning may occur even at temperatures where carbides cannot possibly form. The phenomenon becomes more pronounced in the presence of increasing amounts of martensite, and this accounts for the fact that multiple tempering converts more retained austenite than single tempering for an equal length of time. ●

Sprayed Metal Coatings — Their Structure, Properties and Uses

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Like powder-metal compacts, sprayed metal coatings are porous. An understanding of the effect of this porosity on the properties and limitations of the sprayed metal is essential to a proper use of the versatile metal-spraying process in engineering applications. Included here are some new data concerning the hardness and coefficient of friction of carbon steels sprayed by the wire-gun method, and some examples of the limitations and advantages of metal spraying in production manufacturing.

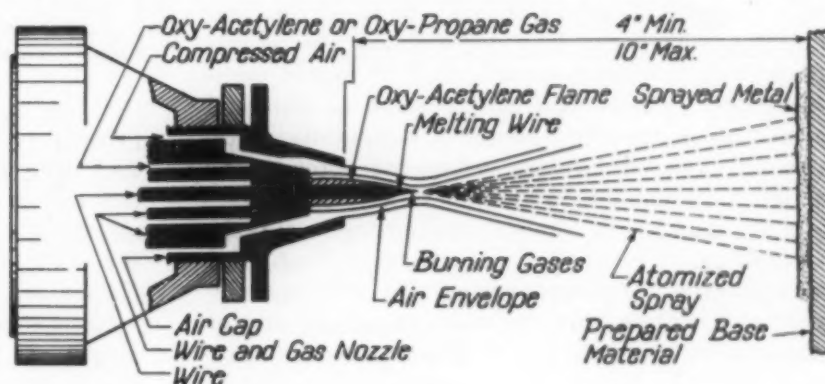
SEVERAL DIFFERENT METHODS have been used for spraying metal coatings by means of compressed air. The most widely used method — and the only one discussed here — employs the so-called "wire gun". Figure 1 shows the action at and beyond the gas head of a wire-gun metal-spraying unit. Wire is moved into the head continuously by means of an air-activated turbine. As the wire passes into the center of the circular flame, it is melted and the molten metal is sheared off by the hot gases surrounding it.

Metal leaves the gun in the form of minute molten globules or spheroids. All but 5 to 10% of them are still molten when they strike the work. Their heat and force cause them to flatten out, almost to spatter on the base.

Because the work is at room temperature when metal spraying is started and each particle is surrounded by a large volume of cool air, the molten globules solidify almost instantly. This instantaneous cooling precludes any significant amount of alloying of the sprayed metal to the base. Only chemical adhesion and mechanical anchorage can cause the sprayed metals to stick to the base.

On nonmetallic bases such as wood, plastics and fibrous materials, adhesion is necessarily mechanical. However, on metallic bases chemical forces probably have some influence in the bonding. During the time of flight of a molten spheroid (about 0.001 to 0.01 sec.), a thin film of oxide is formed on the liquid metal. When the globule strikes the work and flattens, its surface area is greatly increased; the thin film is ruptured; and unoxidized metal is exposed. It is well known that some metals and their oxides adhere very tenaciously to each other. The same type of adhesion may be a factor in the bonding of the slightly oxidized, sprayed particles to a metal base.

Fig. 1 — Section Through the Nozzle of a Wire-Type Metal-Spraying Gun



Most early research on metal spraying was directed toward the improvement of equipment and techniques. In recent years, however, the sprayed metals themselves have been investigated. The sprayed deposit consists of the flattened particles of metal, plus oxides and voids. The properties of the deposit are influenced significantly by both the oxides and the voids.

Porosity of Sprayed Metal

The most obvious property of sprayed metals is porosity. Density (specific gravity) is an inverse measure of porosity. Table I lists the densities of some common sprayed metals and the percentage densities referred to the same metals in the original wire form.

Table I—Density of Sprayed Metals

SPRAYED METAL	DENSITY	% OF WIRE
Aluminum	2.41	90
Zinc	6.36	89
Tin	6.43	88
Tin-base babbitt	6.67	87
Copper	7.53	84
Yellow brass	7.44	88
Commercial bronze	7.57	86
Manganese bronze	7.26	85
Phosphor bronze	7.68	86
Nickel	7.55	85
Monel	7.67	87
Iron	6.72	85
0.10% carbon steel	6.67	85
0.25% carbon steel	6.78	86
0.80% carbon steel	6.36	81
18-8 stainless steel	6.93	88
12% chromium steel	6.74	87

Porosity is both an advantage and a disadvantage. For protection against corrosion, porosity limits metal spraying to metals that are anodic to the base. Thus zinc and aluminum, which are high in the electrochemical series, protect iron and steel in spite of slight porosity. Because zinc and aluminum are soft, their coatings are less porous than those produced from metals of higher melting point. Metals that are cathodic to iron and steel—or some other base—are virtually eliminated from consideration unless the pores in the sprayed coating can be sealed.

In mechanical applications, porosity may be advantageous, particularly for surfaces that must be lubricated. Sprayed metals run with appreciably less friction than the same metals in cast or wrought form. There are two reasons for this: (a) The irregularities in a sprayed metal surface

Table II—Coefficients of Friction of Sprayed Steel

LOAD, PSI.	PLAIN OIL		COLLOIDAL- GRAPHITED OIL	
	HARDENED STEEL	SPRAYED STEEL	HARDENED STEEL	SPRAYED STEEL
445 ft. per min.				
250	0.0107	0.0030	—	0.0021
300	0.00635	—	0.00255	—
400	—	0.0095	—	0.00675
500	0.160	—	—	—
550	0.0201	0.0124	0.0200	0.0097
700	—	0.0152	—	0.0117
750	—	—	0.0248	—
800	0.0224	—	—	—
900	0.0286	0.0174	—	0.0132
1000	0.02175	0.01865	—	0.0136
1050	—	—	0.0236	—
1200	—	—	0.02375	—
1250	—	0.0184	—	0.0136
1300	—	—	0.0234	—
1400	—	0.0174	—	0.0128
1600	—	0.01785	—	0.0120
1850	—	0.0183	—	0.0112
2000	—	0.0179	—	0.0103
261 ft. per min.				
200	—	0.0067	—	0.0057
250	0.0061	—	0.0076	—
550	0.01315	0.0111	0.01175	0.0098
750	—	0.0147	—	—
800	0.0169	—	—	—
850	—	—	0.0112	0.0102
1000	0.0183	0.0149	—	—
1150	—	—	0.0093	0.0078
1250	0.0176	0.01525	—	—
1450	—	—	0.00865	0.0061
1500	—	0.0140	—	—
1550	0.0160	—	—	—
1750	—	0.0152	—	—
1800	0.0156	—	0.0074	0.0039
2000	0.0153	0.01505	0.00745	0.0029
27 ft. per min.				
200	0.0102	0.00745	0.0088	0.0057
550	0.02485	0.0195	0.0209	0.0175
800	0.02475	0.0182	0.0202	0.0156
1000	0.0276	0.01875	0.0222	0.0162
1300	0.0318	0.0200	0.02525	0.0169
1500	0.03125	0.02025	0.0250	0.0161
1750	0.0331	0.02125	0.0265	0.0147
2000	0.0352	0.0220	0.0292	0.0143

provide spaces for the metal particles that are inevitably dislodged during the wearing-in process, and (b) the pores absorb and hold appreciable quantities of oil. Table II shows a comparison of the coefficients of friction of sprayed steel (1.2% C) and of hardened wrought steel (1% C) for various loads and speeds and under two different conditions of lubrication. In these friction tests, the steels were run against a tin babbitt.

Table III—Tensile Properties of Sprayed Metal

METAL	TENSILE STRENGTH, PSI.	STRAIN, % INCREASE IN LENGTH
18-8 stainless steel	30,000	0.27
12% chromium steel	40,000	0.50
Iron	28,000	0.25
0.10% carbon steel	30,000	0.30
0.25% carbon steel	34,700	0.46
0.80% carbon steel	27,500	0.42
Aluminum	19,500	0.23
Aluminum + 6% Si	37,000	0.54
Phosphor bronze	18,000	0.35
Zinc	13,000	1.43

Tensile Strength

In the earliest attempts at testing the strength and ductility of sprayed metals, bases of known strength were sprayed with a coating and the composite piece was tested. Results were highly variable and of doubtful significance. More recent tests have been made in quite a different way so as to measure the deposited metal itself. A coating of predetermined thickness is sprayed onto a bar, and the bar is then pressed out, leaving a tube of sprayed metal. Then the tube is pulled in the tensile test. This was essentially the method used to obtain the values of strength and elongation listed in Table III.

As might be expected of an aggregate consisting of metal, oxides and voids, the tensile strength of sprayed metal is generally lower than that of cast or wrought metal. However, it is higher than was at first believed. With sprayed metal, the tensile test measures particle adhesion almost exclusively, and the structure of the coatings necessarily limits elongation.

It will be noted from Table III that strengths for sprayed steels vary from 27,000 psi. for 0.80% carbon steel to 40,000 psi. for a 12% chromium stainless steel similar to Type 420. The alloy containing 94% aluminum and 6% silicon is stronger in the sprayed state than in other forms. For example, this aluminum alloy sand-cast has a tensile strength of 20,000 psi.; die-cast, 30,000 psi.; but when sprayed, 37,000 psi.

The strength of sprayed metal may be so low as to prohibit its use in some applications. It would, for example, be entirely unsatisfactory to cut threads in a sprayed coating. Similarly, sprayed metal cannot be used as a race for ball or roller bearings. In most mechanical applications,

the base must provide the strength of the part; a thin coating of sprayed metal on the wearing surface will add virtually nothing to the strength. Except in the aircraft industry, where overload and wear allowances must be held to a minimum because of the need for minimizing weight, wear allowances are usually more than ample to permit surfacing or resurfacing with sprayed metal.

Hardness

With sprayed metals, indentation hardness tests like Rockwell measure principally porosity and to only a limited extent the hardness of the sprayed particles. In this respect, sprayed metals are comparable with sintered compacts made from metal powders.

Hardness tests, of course, are useful for comparison of sprayed metals with each other. Table IV presents such a comparison. The values listed should not be interpreted as criteria of the wear resistance of the metals. Sprayed 0.80% carbon steel, for example, though testing Rockwell C-37, will invariably outwear wrought 0.80% carbon steel that has been hardened and tempered to Rockwell C-50. The 12% chromium stainless steel, sprayed on pump sleeves, rods and cranks, has a hardness of only Rockwell C-31 to C-33; however, it cannot be machined readily—not even with carbide tools.

The high resistance to wear of most sprayed metals can be attributed mainly to porosity.

Table IV—Rockwell Hardness of Sprayed Metals

METAL	ROCKWELL HARDNESS	METAL	ROCKWELL HARDNESS
Aluminum	H-72	Tobin bronze	F-78
Zinc	H-46	Nickel	F-87
Tin	H-10	Monel	F-78
Tin-base babbitt	H-58	Iron	B-80
Lead-base babbitt	H-11	0.10% carbon steel	B-90
Copper	F-78	0.25% carbon steel	B-94
Commercial bronze	F-68	0.80% carbon steel	C-38
Manganese bronze	F-76	18-8 stainless steel	B-79
Phosphor bronze	F-75	12% chromium steel	C-32

Porosity causes a decreased contact area and permits the sprayed metal to absorb and hold oil. These two qualities reduce friction appreciably. The hardness of the sprayed metal also influences friction.

Hardness Variations—The hardness of sprayed metal coatings depends on the spraying technique. For instance, variations from a normal flame setting (ratio of oxygen to the fuel gas)

affect hardness, particularly in steels, as shown in Fig. 2. Similarly, varying the distance at which the gun is held from the workpiece (usually from 5 to 6 in.) will affect the hardness of some sprayed metals. Figure 3 shows typical curves for the 0.10, 0.25, and 0.80% carbon steels. Variations in the distance that the wire is extended beyond the gun for melting also affect the hardness. Normally, the melting tip is visible from

rims or tires, or on roller or ball bearing races. Similarly, roll surfaces, where the load is extremely heavy as in steel-mill rolls, are not a suitable base for the coatings. However, the rolls used in some other industries, such as paper, textile and printing plants, have no such excessive loads to bear. In these industries and some others, rolls can be sprayed with stainless steel, bronze or monel to considerable advantage.

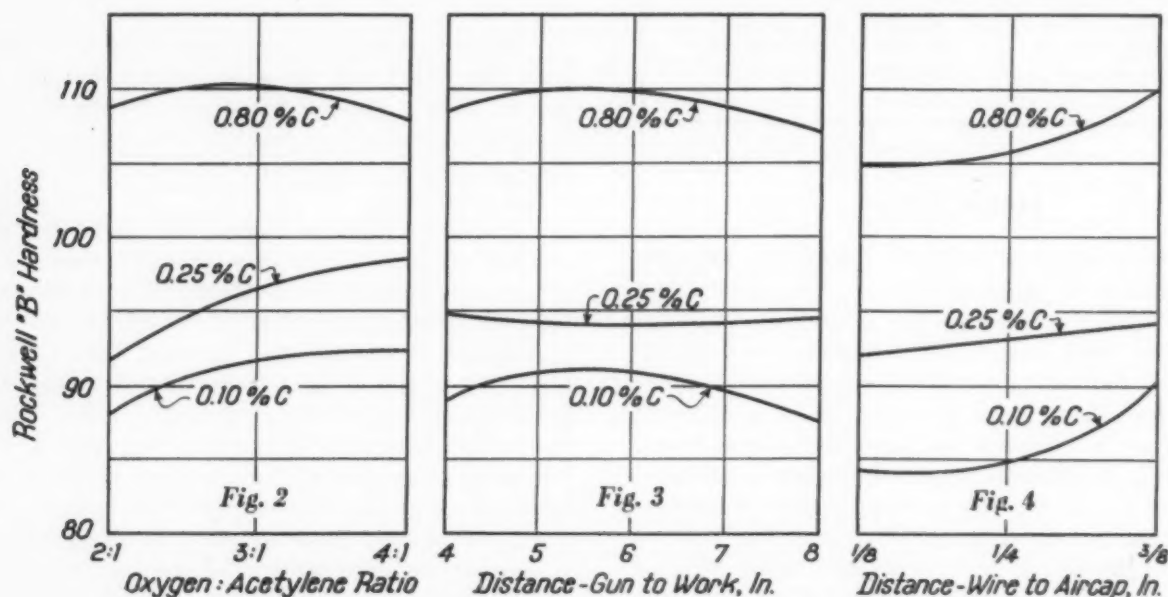


Fig. 2, 3 and 4 — Effects of Variations in Spraying Technique on the Hardness of Sprayed Steel

$\frac{3}{16}$ to $\frac{3}{8}$ in. beyond the barrel of the gun, depending on the metal being sprayed. Figure 4 indicates that the hardness of sprayed carbon steels increases as the wire is melted off farther and farther outside the gun.

Limitations and Advantages

Inherent in the structure and properties of sprayed metals are both the limitations and the advantages of metal spraying as a process. Some of both have been mentioned in the discussion of individual properties; they are summarized here with specific examples.

Limitations—Because of low elongation and the type of bond, sprayed metals should not be used where they will be subjected to direct impact. There are very few surfaces in machines where such a condition of impact exists in service; one such place is a valve seat in an internal combustion engine.

Low elongation and reduced tensile strength do not permit sprayed metals to withstand rolling action. Hence, they should not be used on wheel

Cathodic coatings have already been mentioned as being unsatisfactory in many instances. However, lead and tin coatings can be used for some purposes, because they are relatively soft and can be sealed by shot-blasting at reduced blast pressures, using round steel shot. The blasting pressure is usually about half that commonly used for other types of blasting.

Advantages—The largest single advantage in spraying metals appears to be the versatility of the process: Any metal that can be made in wire form and melted in the flame can be sprayed,* and the selection of base materials that will accept sprayed metal is almost unlimited. Wood, paper, plaster, cloth and fibrous materials can be coated. Furthermore, these nonmetallic bases generally need no preparation except thorough cleaning. Plastics, glass and ceramics will also take sprayed coatings. For metal bases, there is almost no limit, because metals can be prepared to receive the coating in several different ways, depending on operating

* **EDITOR'S NOTE:** The powder method of spraying (not discussed in this article) is particularly suitable for metals not readily produced in the form of wire.

conditions, loads, thickness of coating desired and the mechanical properties of the base itself.

Second in importance is the speed of modern metal-spraying guns. The unit illustrated in Fig. 5, for example, melts and sprays $\frac{3}{16}$ -in. wire and attains practical speeds of deposition varying from 20 lb. per hr. of carbon steel or stainless steel to about 130 lb. per hr. of lead.

The third advantage of metal spraying is the low temperature of the work during the process. It has already been pointed out that the particles of metal are extremely small, as sprayed, and are surrounded by cool air. As a result of this, the temperature of the base material is seldom raised above 200° F., and when desired, can usually be kept well below that temperature.

The reduced friction of sprayed metal coatings has been discussed in preceding paragraphs.

Applications

Present production applications fall into three main categories—electrical, protective and mechanical.

Electrical—Probably the largest single application for metal spraying in the electrical field is the spraying of carbon products with copper. Carbon needs no preparation; sprayed copper adheres to it readily. The sprayed copper is excel-

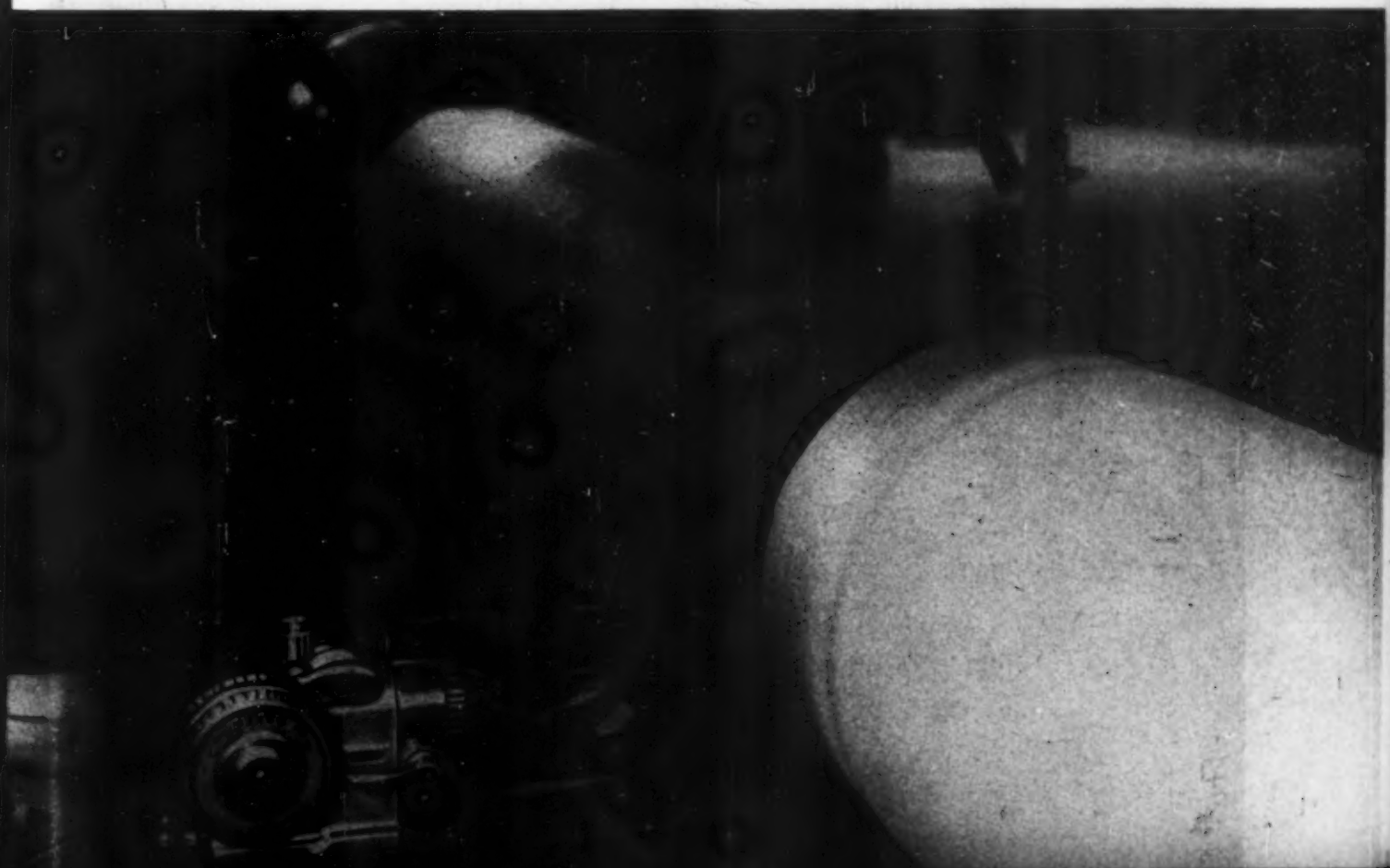
Fig. 6 — Radio Circuits Sprayed Through a Stencil

lent for contact purposes and also permits the soldering of metallic conductors to the carbon.

Another large application in this field is the use of aluminum for condensers, resistors and the like. Here, the aluminum is sprayed onto cloth or paper, or some other fibrous material.

Electrical circuits are being sprayed on glass, ceramics and plastics of several kinds. Figure 6 shows radio circuits sprayed onto a dielectric material through a stencil. A particularly promising technique for this application involves the molding of grooves in the panel before spraying.

Fig. 5 — Wire-Type Metal-Spraying Unit Mounted on the Tool Post of a Lathe. This type of gun is used for production applications where many pieces or large surface areas of one piece are conveyed past the spraying unit.



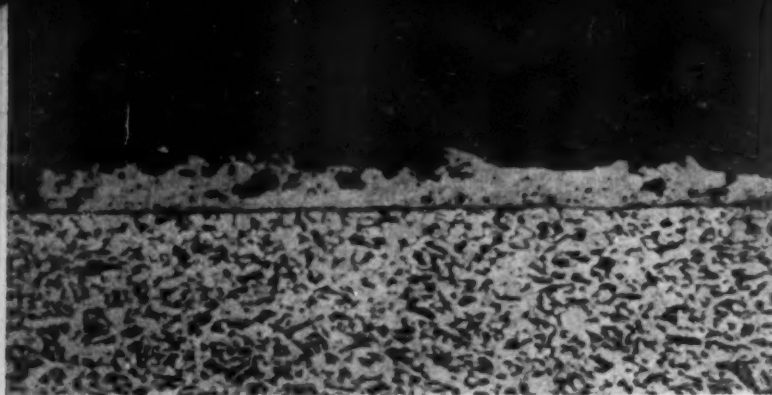


Fig. 7 — Mild Steel Base (bottom) Sprayed With Undercoat (white central band) and 0.10% Carbon Steel (top). 100×

Mechanical—The possibilities of sprayed metal for the production of mechanical parts seems to have lagged behind other applications. However, one particular example concerning a turbine shaft will indicate a type of application that might be investigated more fully. On this turbine shaft, there are only four surfaces that are subject to the combination of wear and corrosion. To use stainless steel for the entire shaft would be expensive, both for material and machining. These costs were reduced substantially by using mild steel for the shaft and protecting the critical areas with sprayed stainless steel. With the particularly favorable lubricating qualities of sprayed metals, a similar selective application might be feasible for shafts subject to excessive wear, or shafts that are difficult to lubricate.

One other interesting development is worthy of mention in connection with the mechanical field. Out of present research has come the discovery of a material that, when sprayed, bonds itself to most metals without any preparation

except cleaning. It has already been used as a bonding undercoat on many mechanical parts (Fig. 7). The importance of this undercoat is that it reduces greatly the amount of mechanical preparation recommended heretofore.

This new material (containing more than 50% molybdenum) bonds readily to magnesium and aluminum. Hence, it opens up the intriguing possibility of using the light metals for parts that have not been practical in the past. For example, it is possible to apply the bonding undercoat to magnesium and then to apply a coating of 0.004 to 0.005 in. of copper, on which soldering may be done. Or a hard, wear resistant metal such as 12% chromium stainless steel can be sprayed over the undercoat on magnesium or aluminum.

Protective—In the field of corrosion protection, particularly of iron and steel products, there are a number of practical possibilities, mainly for zinc and aluminum. Protection of the finned barrels of aircraft engine cylinders is the best known application of this kind. At the peak of production, hundreds of thousands of pounds of aluminum were sprayed for this purpose.

Figure 8 illustrates another protective job on which metal spraying has proved effective. Steel caps for safety shoes are sprayed for two reasons. One, of course, is protection against corrosion. The other is to prevent sparking if a metal object should strike the steel cap. Around gas plants, oil refineries and powder operations, this is important.

Speed nuts, lock nuts, and such parts have been metal sprayed in tremendous quantities. ☉

Fig. 8 — Commercial Bronze Being Sprayed Onto Steel Caps for Safety Shoes



Some Industrial Uses of Nitrogen and the Rare Gases

By J. M. Crockett

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Nitrogen, helium and argon are being used increasingly for flushing hydrogen from liquid metals and for protective atmospheres in heat treating. In inert-gas-shielded arc welding, argon or helium is used to exclude the oxidizing air from the electrode, the arc and the weld puddle. All of these applications are discussed in the present article.

THE INDUSTRIAL APPLICATIONS of nitrogen and the rare gases are based on the chemical inertness of these gases. Although nitrogen is not an inert gas under all conditions, it is used where oxidizing or reducing gases would be detrimental. Processes utilizing argon, helium or nitrogen are varied, and the applications of these gases are growing. This growth can be illustrated best by considering some of the recent developments in metal refining, welding, and protective atmospheres.

Metal Refining

It is often difficult to produce porosity-free castings from certain metals and alloys because of hydrogen dissolved in the molten metal. Figure 1 illustrates the solubility of hydrogen in aluminum, iron and nickel.* If, during melting and refining, more hydrogen is dissolved by the molten metal than the solid metal will hold, the excess hydrogen will be liberated as the metal cools. Thus, the

*EDITOR'S NOTE: Recent data concerning the solubility of hydrogen in liquid and solid aluminum and some considerations regarding the flushing of liquid aluminum are reviewed in the abstract on p. 892.

hydrogen gas may become trapped in the solidifying metal, causing porosity in the casting.

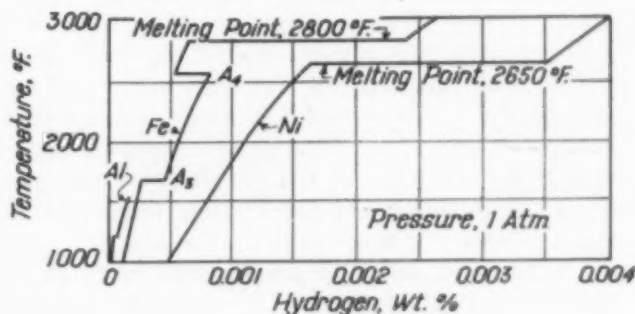
There are several means of eliminating dissolved hydrogen from molten metals. One method that is gaining increasing recognition is flushing the molten metal with a dry, inert gas in the furnace, before casting.

The flushing process consists of bubbling an inert gas slowly through the molten metal. The action of the flushing gas in removing the dissolved hydrogen is purely physical. The rising bubble of inert gas as it is liberated from the flushing tube is free from hydrogen.

Since there is a definite pressure of hydrogen in the molten metal, the hydrogen diffuses from the metal into the inert gas bubble, and is carried from the metal and released into the atmosphere when the bubble breaks through the surface of the bath. Thus, the hydrogen is literally swept from the bath. The inert flushing gas does not dissolve in the metal, but passes off harmlessly.

The gases generally used for flushing are nitrogen and argon. Because nitrogen is cheaper

Fig. 1 — Solubility of Hydrogen as a Function of Temperature in Aluminum, Iron and Nickel



and more readily available, it is used wherever it will not react with the metal or with one of the alloying constituents.

The equipment required for flushing is simple; only a regulator, a hose, and an injection pipe are necessary. The flushing action in molten aluminum is shown in Fig. 2. The nitrogen is injected at a low pressure, so as to produce a gentle bubbling

is in developing suitable injection devices that will not be dissolved by the steel bath. Theoretical considerations of the flushing process indicate that the smaller the bubble size the greater the efficiency. Therefore, if proper means for diffusing the flushing gas in molten steels can be obtained, certain other defects, such as flaking, which are also attributed to dissolved hydrogen, may be eliminated.



Fig. 2 — Flushing Aluminum With Nitrogen in a Crucible. The flushing tube is in the center.

action. In large plants the gas is supplied from manifolds or bulk-delivery units and piped to the various furnaces. In addition to removing dissolved gas and preventing porosity in aluminum castings, flushing with nitrogen floats nonmetallic particles to the surface of the metal and thus helps to produce clean aluminum.

Figure 3 shows samples cast from a heat of stainless steel. The porosity resulting from dissolved hydrogen is conspicuous in the casting on the left. By flushing the heat with argon, enough hydrogen was removed so that sound castings with normal shrinkage were produced, as shown in the sample on the right in Fig. 3. Because nitrogen gas may react with steel at refining temperatures, argon is used for flushing.

Laboratory and field development work in the flushing process is active. The principal problem

Nitrogen is also being used to some extent in refining secondary aluminum. When magnesium is removed commercially from molten aluminum, a chemical flux is used which reacts with the magnesium. This flux may be shoveled onto the molten bath and rabbled with hand tools, or it may be injected into the bath, using nitrogen as a carrier. The advantage of using nitrogen is that a faster reaction is obtained between the flux and magnesium. In addition, the agitation of the nitrogen keeps the furnace refractory more nearly free from dross, and thus prolongs the furnace life.

Inert-Gas-Shielded Arc Welding

At the high temperatures in unshielded arc welding, the metals being welded may react with the atmosphere surrounding the heated area. This

may cause undesirable inclusions in the weld, and surface discoloration.

The welding of some aluminum and magnesium alloys by the conventional metal-arc process has been difficult or impossible because of the high affinity of the metal for oxygen or nitrogen in the air. This difficulty has been lessened by using chemical coatings on the electrodes that provide some shielding when the electrode is melted in the arc. These coatings also form slag that helps to remove undesired inclusions from the weld and provides a protective layer over the hot welded area.

Although metal-arc welding is, in general, still the best process for joining mild steel, stainless steel and many cop-

per and nickel alloys, a comparatively new welding process that uses an inert gas shield supplements the older process and widens the field of application.

In inert-gas-shielded arc welding,* the heat of fusion is derived from an arc between a non-

*EDITOR'S NOTE: Two other, newer welding processes that use inert gas for shielding are described in the abstracts on p. 886 to 890 in this issue. One of these new processes, inert-gas-shielded metal-arc welding, is similar to conventional gas-shielded arc welding but uses a consumable filler wire instead of a non-consumable tungsten electrode. The other new process, inert-gas-shielded-arc spot welding, produces a spot weld by fusion only, from one side of the joint.



Fig. 3 — Samples Cast From a Heat of Stainless Steel. The sample at the left was taken before the molten metal was flushed. The one on the right was taken from the same heat after the liquid metal was flushed with argon.

magnesium. These metals are difficult to weld because of the refractory oxides that form during conventional welding and because the use of special chemical coatings or fluxes is necessary to obtain satisfactory welds. Welds made in these metals by the shielded-arc process are made without flux, since the inert gas (argon or helium) protects the molten metal from oxidation.†

Stainless steels and copper and nickel alloys are also being joined successfully by inert-gas-

†EDITOR'S NOTE: For a report concerning the availability and use of high-purity helium (99.8% pure) for the welding of aluminum, see p. 848 in this issue.

consumable electrode (usually tungsten) and the work; and the electrode, the arc and the weld area are shrouded by an inert atmosphere (either argon or helium) supplied through the electrode holder (Fig. 4 and 5). Argon and helium are completely inert and give excellent protection to the weld area; the resulting welds are therefore free from inclusions. No flux is necessary, and there is no slag to remove or to become trapped in the solidifying weld.

Inert-gas-shielded arc welding not only protects the metal during welding but also gives welds of better appearance and permits faster welding. Inert-gas-shielded arc welding is popular for welding aluminum and

Fig. 4 — Water-Cooled Electrode Holder for Use in Manual Welding by the Inert-Gas-Shielded Arc Method

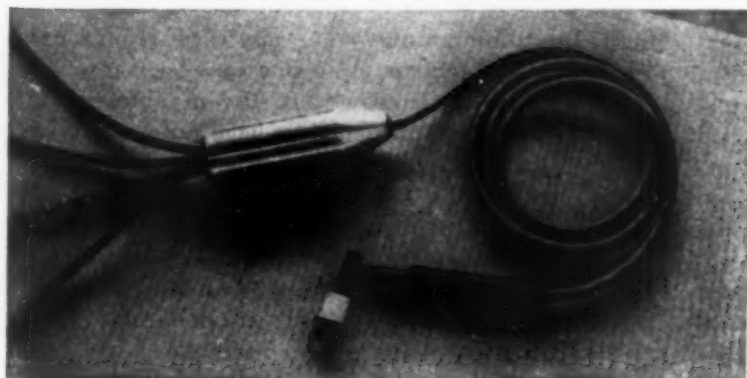
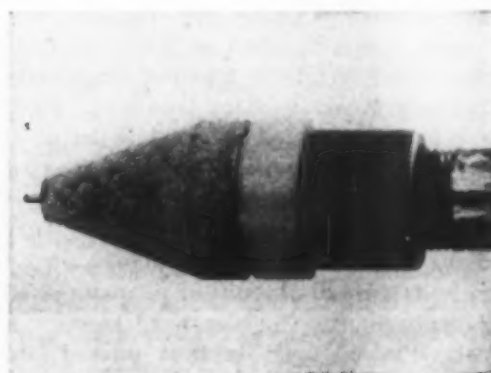


Fig. 5 — Ceramic Cap Around Electrode to Direct the Flow of the Shielding Gas



shielded arc welding, without the use of fluxes. At present, the principal application for gas-shielded welding with these metals is on material thinner than about $\frac{3}{16}$ in., for which the speed of welding is exceptionally high. Generally the cost of finishing (grinding or machining) the weld is less because the area involved is smaller than when the weld is made by some other process.

Inert-gas-shielded arc welding is still a relatively new process and its usefulness is expanding.

Protective Atmospheres

Protective atmospheres are used to minimize or prevent oxidation and to exclude undesirable gases from material during processing. Obviously these applications are most widespread in the chemical and metal industries, and the results desired will dictate the gas and purity of the gas to be used.

Although for many purposes an atmosphere produced by burning a fuel gas with a limited or controlled supply of air is completely satisfactory, special products require gas compositions that can be met consistently only by using the rare gases or commercial nitrogen as the protective atmosphere. For example, in heat treating steel, it is often necessary to prevent surface oxidation or decarburization, or both, during the heat treating cycle. Commercial nitrogen is now being used in several large-scale operations to achieve results that could not be obtained by the more common protective atmospheres, such as combusted fuel gas (purified or unpurified), reacted fuel gas, or dissociated ammonia.

High-Carbon Steel—The chemistry of protective carbonaceous atmospheres is complex, particularly when steel is the metal being heated. It is not enough to provide an inert atmosphere, for a steel heat treated in pure nitrogen will oxidize some because the adsorbed oxygen present on the surface of the steel will react with the metal during the heat treating cycle. Consequently, when nitrogen is being used as the basis of a heat treating atmosphere, it is usually desirable to add a small amount of active reducing gas. One example of such a modification of nitrogen is the atmosphere used for heat treating high-carbon steels, especially those used for bearings. Here the steel must be heat treated without depleting the carbon content of the surface, for if decarburization occurs, the surface of the bar or rod will be too soft and will have to be ground or further processed to remove the unsatisfactory surface. Nitrogen with a small amount of propane is an excellent atmosphere for high-carbon steels. For other steels, the nitrogen is first passed through hot

charcoal to convert the residual oxygen to carbon monoxide. By use of these nitrogen-base atmospheres, economies in final surface preparation of high-carbon steel may offset the higher initial cost of the nitrogen, as compared with generated atmospheres.

Cold Rolled Steel Strip for Tin-Plate—The composition of the atmosphere used in the bright annealing of cold rolled steel strip is one of the important factors in producing electrolytic tin-plate. Electrolytic tin-plate made from steel that has been annealed in an atmosphere of nitrogen and a few per cent hydrogen is superior to electrolytic tin-plate produced from strip annealed in the conventional atmosphere generated from fuel gas. Although the exact reason for this difference is not known, the fact has been confirmed in the laboratory and by plant operations.

The nitrogen-base annealing atmospheres have an added advantage, because the composition can be controlled merely by adding the desired amounts of active reducing gas. The result of using such a mixture is that the amount of oxygen or oxidizing compounds in the gas is very low. Therefore, the concentration of active reducing constituents need be only high enough to react with adsorbed oxygen or with oxygen that may diffuse into the furnace. With active reducing gases held below 2 or 3%, the gas mixture is safe and nonexplosive and can be used for the whole cycle.

The close control of gas composition becomes important where long annealing cycles are used. The desired surface condition is achieved by adding to nitrogen fixed volumes of active reducing gases, such as hydrogen or propane. This positive control of the entering gas, where the purpose of the atmosphere is to prevent oxidation or decarburization, is desirable and eliminates the necessity of checking the gas in the furnace itself by gas analysis for control. However, in applications where carbon is added to the steel by means of the atmosphere, as in gas carburizing, it is desirable to use generated gas with high percentages of active reducing gases to facilitate control through gas analysis.

Ductile Titanium—The rare gases are not used much as protective atmospheres except for laboratory work. However, in the manufacture of ductile titanium pure helium is often used as an atmosphere. Titanium metal reacts to form chemical compounds with both oxygen and nitrogen and readily dissolves hydrogen at elevated temperatures. Consequently, the reduction of titanium compounds to titanium metal, and the subsequent treatment of the metal powder are done in a helium or argon atmosphere or in a vacuum. ©

A Mechanical Test for Detecting Longitudinal Fissures in Fine Wire

By D. W. White*

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This test, which the author has named the "knife-edge" test, determines the ability of a wire to withstand tension over a relatively sharp knife edge of known radius of curvature while being rotated about its longitudinal axis. Straining the specimen in this way will cause fracture of a fissured wire but not of a flawless one. The test is being used in the factory for the quality control inspection of semifinished tungsten wire for lamp filaments. It should be applicable to other wire also.

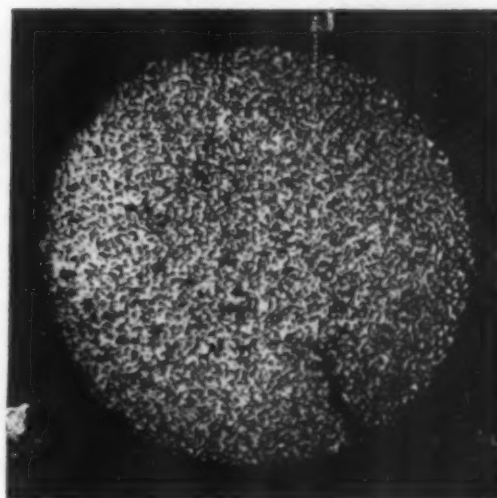


Fig. 1 — Micrograph of a 0.003-In. Tungsten Wire Containing a Longitudinal Fissure

FOR MANY USES wire must be free from longitudinal cracks, seams or fissures. This is particularly true of fine tungsten wire used in the making of lamp and vacuum tube filaments, in which stresses other than pure tension are likely to exist during processing and operation. Longitudinal fissures such as shown in Fig. 1 inevitably cause premature failure of the filament.

Heretofore there has been no feasible test by which wire containing fissures could be segregated from wire that was free from such defects. Conse-

quently, fissured wire, as well as flawless wire, has been processed into finished or semifinished products. This unavoidable practice gave high rejection losses; the defective wire went through useless processing operations; and other materials in the product assembly were wasted.

The presence of longitudinal fissures in a wire cannot be detected by conventional tensile testing; the yield and tensile strengths are not changed by the presence of fissures, because the fissures run parallel to the direction of stress, and the cross-sectional area is effectively the same for fissured wire as for flawless wire of the same nominal diameter. Torsion, torsion-tensile and torsion-fatigue tests have also been ineffective. Various magnetic, sonic and electronic devices, which have been employed commercially for the inspection of heavy wire, rods and bars, apparently have not yet been used satisfactorily in locating fissures in fine wire.

The only method that has been even moder-

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Dr. R. M. Zabel, formerly general engineering manager of the Lighting Division of Sylvania Electric Products, Inc., now of Westinghouse Electric Corp., contributed many constructive suggestions in connection with the development of the test described here. The author is also grateful to Mr. W. E. Kingston, manager of metallurgical research and development, Sylvania Electric Products, Inc., and to other persons in that company, for suggestions and data.

ately successful in determining whether a sample of wire contained fissures has been the metallographic examination of cross sections. However, this method has several disadvantages.

1. It requires metallographic equipment and a microscope, which represent a high initial investment if used for no other purpose.

2. It is time-consuming and tedious.

3. A skilled operator is required.

4. Operating costs are relatively high, not only for labor but also for materials, such as the bakelite or other mounting material.

5. The percentage sampling is extremely small because it is too laborious to check the wire at more than a few points in a spool of wire that might be several thousand feet long. If the fissures were intermittent, rather than continuous, they could easily be missed.

Therefore, it has been feasible to use the metallographic method only infrequently, when quality demands were great enough to warrant the expense, time and effort of such an inspection.

The inspection of wire for fissures by the test described herein, which we shall call the "knife-edge test", overcomes the disadvantages of metallographic examination. It is rapid, simple, dependable, inexpensive initially and in operation, relatively foolproof, and it tests the wire continuously along the length of a sample.

The Knife-Edge Test

If fissured tungsten wire is bent sharply, then straightened and pulled in tension, perhaps 5 to 10% of the specimens will give low values of tensile strength (as low as 15% of the normal tensile strength), whereas flawless wire, when subjected to this same treatment — bending, straightening and tensioning — will be consistently high in tensile strength (about 80 to 90% of the normal tensile strength). This result demonstrates that there is a certain critical orientation of the fissure plane with respect to the axis of the bend for which the wire will be extremely weak when restraightened and pulled in tension. Such an explanation accounts for the fact that fissured wire is only occasionally weak — that is, in only a minority of the specimens does the bend fulfill the conditions necessary for subsequent weakness.

The knife-edge test is a practical, refined method for carrying out the essential steps of bending, straightening and applying tension. The test determines the ability of a wire sample to withstand pulling under controlled tension over a relatively sharp edge of known radius of curvature while being rotated about its longitudinal axis.

A setup used by Sylvania Electric Products,

Inc., is illustrated schematically in Fig. 2 and photographically in Fig. 3. In testing a sample of wire, one end of the sample is fastened to the weight assembly of specified mass. The wire is then extended vertically upward and over the knife edge of specified radius and then horizontally to a motor-driven shaft, where the other end of the wire is passed around a pin and fastened to a clip. It has been found desirable to have the weight assembly resting on its side on the floor or other suitable support, so that when the test is started the load is applied gradually at first, thereby avoiding a sudden impact load on the wire. When the motor switch is turned on, the

wire is wound around the rotating shaft, causing the weight assembly to rise and the portion of the sample that was initially below the knife edge to pass continuously up and over the knife edge at

an angle of 90°. Simultaneously with this movement, the wire also rotates about its longitudinal axis by drifting downward along the knife edge, which is deliberately inclined. The winding shaft is also sloped at the same angle with, and parallel to, the knife edge so that the wire is always being bent very nearly 90°. The desirable angle of inclination is the one that will allow each wire sample to travel nearly the full length of the knife edge and no farther. The motor is operated until the wire fails, revealing the presence of a fissure, or until the weight approaches the knife edge, proving the wire to be free from fissures.

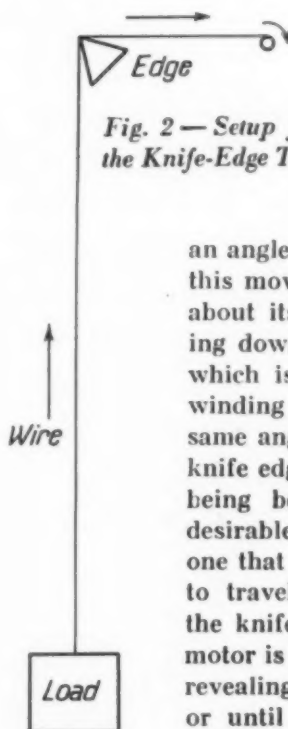


Fig. 2 — Setup for the Knife-Edge Test

The knife edges are prepared from tungsten carbide bars 3 in. long and having the cross section of an equilateral triangle $\frac{1}{2}$ in. on each side. The triangular cross section permits the wire to contact only the edge and to escape the faces of the bar. Each bar accommodates three different edge radii, which are machine-ground and polished, using a special jig. It is important that the edges have a smooth surface finish so as to avoid galling and digging the wire. In the tester, the carbide bar is held rigidly in a cradle support with one of its edges exposed. The bar is easily removable for changing edges.

The winding shaft is $\frac{5}{16}$ in. in diameter and 4 in. long. It is driven by a universal-type, fractional-horsepower motor coupled to a reduction



Fig. 3 — Photograph of the Knife-Edge Tester in Operation

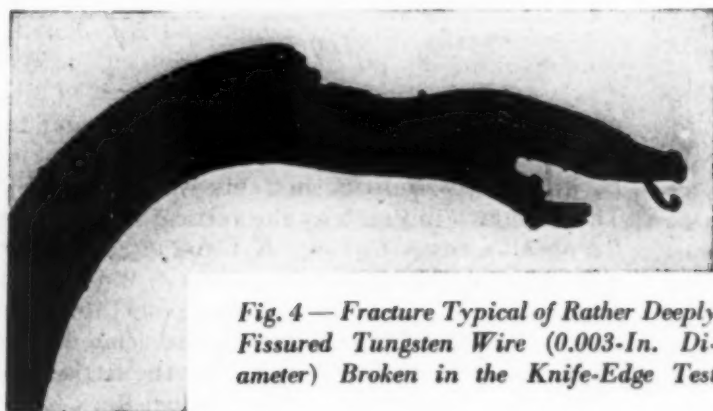


Fig. 4 — Fracture Typical of Rather Deeply Fissured Tungsten Wire (0.003-In. Diameter) Broken in the Knife-Edge Test

gearing unit that delivers a shaft speed of 60 r.p.m. and an output torque of 5 lb-in. The set of weights consists of a jaw clamp attached to a threaded rod on which calibrated, doughnut-shaped disks of varying thicknesses and diameters can be hung, sandwiched between two calibrated nuts. With this simple, convenient weight assembly, various loads can be formed readily and fastened to the wire easily. By various combinations, a range of loads can be covered, preferably in increments of 25 g. or less.

For ease of manipulation, the tester is mounted at about shoulder height. A sample of wire from 4 to 5 ft. long can thus be accommodated for each specimen. The floor directly below the tester is padded so as to absorb the shock of the falling weight when the wire breaks. A shallow box lined with sponge serves this purpose well.

Theory of Operation of the Knife-Edge Test

That the knife-edge test performs the three essential operations — bending, straightening and tensioning — is proved by the fact that a fissured wire breaks not at the knife edge but in the straight portion of the wire between the edge and the winding shaft, after having passed over the edge and having been straightened and pulled. Thus, the failed end of the portion of wire to which the weight is attached has a 90° hook. A "fish-tail" type of break is characteristic of a specimen having a fissure of significant depth. A typical break is shown in Fig. 4. This fracture is distinctly different from the straight-across type of break obtained from brittle wire bent sharply, as in the customary qualitative brittleness test of bending a wire sample over a pronounced edge. Furthermore, a brittle fracture would occur at or very close to the bend.

The wire is caused to drift laterally along the sloped knife edge during testing primarily so that it will rotate many times through every possible orientation of a prospective fissure with respect to the axis of bending, and thus, at some time during the test, the fissure will arrive at the critical position necessary to allow the wire to break after straightening and pulling. This lateral movement of the wire further serves to distribute the wear of the knife edge over its entire length.

The mechanism for the unique behavior of fissured wire under the knife-edge test is believed to be as follows: When the wire is bent sharply, any fissure whose plane is properly oriented with respect to the bending axis is propagated in one or more paths across the wire at the bend, thus dividing the wire into two or more separate strands at this point. The tensile stresses that accompany the bending cause the outer individual strands to elongate plastically more than the inner strands, thus resulting in strands of varying lengths. At the same time, the inner strands are workhardened by the compressive stresses of bending, thus potentially reducing the available longitudinal elongations of these strands. When the wire is then straightened and tensioned, each strand must support the full tensile load individually — first the shortest strand, then the next shortest, and so on. Because each of these individual strands is unable to support the full load,

the wire breaks by the progressive, steplike failure of the strands. Were the wire free from fissures, the stress and strain gradients from tension to compression in bending would be continuous and uninterrupted and, upon straightening, the return to the full tensile loading would be supported by the entire cross section of the wire, with no resultant failure.

Test Factors

The size of the bending radius and the amount of load applied are governed mainly by two factors: (a) the diameter of the wire being tested and (b) the degree of sensitivity desired in the test. For any given degree of sensitivity in the test, the finer the wire, then the sharper the knife edge or the smaller the applied load. For any given wire size, the sensitivity can be increased by decreasing the radius of the knife edge or by increasing the load. (In practice, the sensitivity is usually regulated only by changes in load.) In this way the test can be made sensitive to extremely small longitudinal surface defects, even to include some die marks, or on the other hand, conditions can be adjusted so as to segregate only wire having fissures of a certain minimum depth, depending on the requirements of the particular wire in service.

Testing "Black" Tungsten Wire

Tungsten wire in the as-drawn condition and coated with the graphite wire-drawing lubricant is termed "black". A general method has been established for determining the knife edge and load to be used in testing black tungsten wire of varying sizes at various degrees of test sensitivity. Experimental data for numerous sizes of wire having fissures of comparable proportional depth have indicated that the wire size (W) expressed in terms of weight per unit length* is a function of the square of the optimum knife-edge radius (E_0), when test loads (L) are used that are proportional to the wire sizes: $W = fE_0^2$ when $L \propto W$. By properly choosing the weight and the dimensional units for W and E_0 , f can be made to equal unity, and the equation becomes $W = E_0^2$. This is expressed graphically in Fig. 5 by the dash-dot

*Because micrometer measurements are not sufficiently accurate in the range of fine wire sizes, it is customary to weigh a given length of wire on a torsion balance to determine the wire size, which is then expressed in terms of weight per unit length. When dealing with wire of one material, where density remains relatively constant, wire size can be accurately designated by this type of measurement, and direct comparisons can be made between one lot and another.

line. It assumes, however, that an infinite number of knife edges are available; actually, we must confine ourselves to the limited number of specific edges (E_s) that we have. Thus, in selecting the proper edge for testing wire of a certain size, we

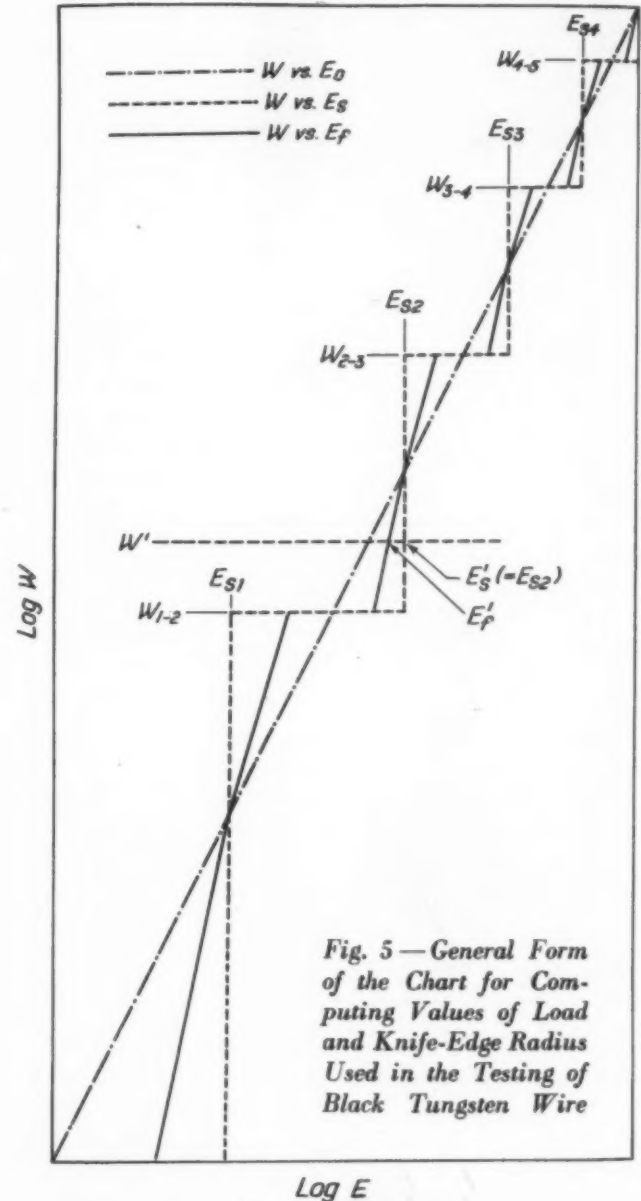


Fig. 5—General Form of the Chart for Computing Values of Load and Knife-Edge Radius Used in the Testing of Black Tungsten Wire

must use the edge with radius nearest in value to the square root of the wire size. That is, E_s will be closest to E_0 in value when $E_0 = \sqrt{W}$. This is shown in Fig. 5 by the vertical dotted lines for specific edges E_{s1} , E_{s2} , E_{s3} , and E_{s4} . (There will obviously be certain wire sizes— W_{1-2} , W_{2-3} , W_{3-4} , and W_{4-5} —with square roots that lie exactly halfway between two neighboring knife-edge radii. These are indicated by the horizontal dotted lines joining the vertical dotted lines. For

such sizes, either of the two knife edges may be used in testing the wire.)

The intersections of the dash-dot line with the vertical dotted lines represent those ideal wire sizes for which the specific knife edges used in testing these sizes are also the optimum knife edges — that is, those wire sizes that are numerically equal to the square of the specific edges that are used. For each available knife edge there is only one such wire size. As previously mentioned, under these optimum conditions the test loads are proportional to these ideal wire sizes and are therefore proportional to the squares of the optimum knife-edge radii: $L \propto E_0^2$.

Now the above relationship would give a direct means for calculating the test load for any size of wire if we had an infinite number of knife edges available and could use the corresponding optimum edge in testing each size of wire. However, for wire sizes other than the ideal sizes referred to in the previous paragraph, we cannot use the E_0 values for calculating loads, since the knife-edge radii available for test are E_n , not E_0 . Neither can we use the E_s values for calculating test loads, because such a procedure would not discriminate between variations in wire sizes that are tested on any single knife edge. Therefore, we must set up fictitious knife-edge values (E_t) for purposes only of calculating test loads. These values of E_t lie midway between the E_s and E_0 values and are represented by the solid lines in Fig. 5. (For wire sizes W_{1-2} , W_{2-3} , W_{3-4} , and W_{4-5} , there are obviously dual E_t values as well as dual E_s values. If the smaller knife-edge radius is selected for the test, then the smaller E_t value should be used for calculating the test load. Conversely the upper value of E_t should be used with the larger edge radius.)

The test load, then, depends on E_t^2 . It should also depend on the tensile strength (S) of the wire, for failure of fissured wire in the test is by tension. Thus, the test load is calculated from the expression: $L = kSE_t^2$. The constant k is merely a sensitivity factor. That is, it can be used to regulate the sensitivity, or rigorosity, of the test, and can be standardized for any particular wire application, depending on the maximum depth of fissure permitted.

This procedure for determining the load and the knife edge to be used in testing black tungsten wire of a given size is simple to use. It is necessary merely to consult a previously standardized chart of the type shown in Fig. 5 and to pick off the E_s and E_t values for the size of wire in question. E_s is the knife edge that should be used in the test. The test load L is obtained from the E_t value, the tensile strength S , and the sensitivity factor k , according to the formula $L = kSE_t^2$.

For example, if we desire to test several samples from a spool of wire of size W' , we find from the chart that the E_t value is E'_t . Assuming that we have measured the tensile strength of this wire and that we customarily use a certain known sensitivity factor for the particular type of wire in question, then we can quickly calculate the test load. We also find from the chart that we should test these samples over knife edge E'_s , which in Fig. 5 is seen to be E_{s2} .

Except for extremely fine wire, the foregoing method of computing values for knife edge and load has proved useful over a wide range of wire sizes. For wires considerably smaller than 0.001 in. in diameter this treatment is not strictly valid, and it should be modified with correction factors, which must be determined experimentally.

Testing Other Types of Wire

The test conditions for clean, annealed tungsten wire differ from those for black wire, because (a) the different surface condition affects the frictional resistance when the wire passes over the knife edge and (b) clean wire is usually weaker than black wire. Furthermore, since clean wire may be more brittle than black wire, the knife-edge test would probably detect variations in brittleness as well as fissures, thus necessitating close observation of the type of wire fracture.

No attempt has yet been made to inspect materials other than tungsten for longitudinal fissures by this method, but it would be expected that the knife-edge test would be suitable for any type of wire having small plastic elongation combined with good ductility (drawability).

Conclusions

The knife-edge test, as described herein, has proved to be a useful tool and a highly dependable method for evaluating the quality of wire, provided the sampling is done correctly and in sufficient amounts, which, of course, is an important factor in any type of destructive testing. This test is now used to great advantage in the factory for the quality control inspection of black tungsten wire for longitudinal fissures. It has replaced the more costly and less reliable metallographic method of examining cross sections and has made it economically feasible to inspect tungsten wire more thoroughly. In this way defective lots of wire have been segregated before being processed into finished products, resulting in less waste of labor and materials, fewer rejections, and higher quality of product. ●

Search for Atomic Control Continues

EXTRACTS from official records of the United Nations Atomic Energy Commission—the committee charged with devising a plan for international control—have been presented in *Metal Progress* from time to time. These outlined the original American proposal and the improvements suggested by the subcommittee of experts, as well as the Russian counter-proposals. By votes of 9 to 2 (the Soviet Union and the Ukraine in the minority) it was decided that adequate control is now technically possible, but politically impossible, and on May 17 the problem was turned back to the General Assembly of the United Nations. The latter referred the problem to its Political and Security Committee, and on Oct. 1 and 2, the Soviet Representative, ANDREI VISHINSKY, restated his government's position when opposing a resolution to approve the work done so far by the U.N. Commission and to suspend further work. Extracts from his speeches follow:

"If we analyze the work of the Atomic Energy Commission we find that the United States Government at the very outset would not agree to eliminate and prohibit atomic weapons, but shifted all its attention to the establishment of international control. However, you cannot control atomic weapons if you have not first prohibited their production.

"Destruction of the atomic bomb is not sufficient. 'A strict and effective international control organ', in the words of Generalissimo Stalin, is indispensable. However, international control under the United States plan would be an American agency, designed as a monopolistic super-trust which would own all atomic facilities and all enterprises connected in any way with the production of atomic energy. The representatives of Eastern Europe can only expect that such an agency will prescribe such regulations as will stifle us, strangle our national economy and bind and tie us hand and foot.

"The American plan says that the control must be in 'successive' stages, unspecified as to time or scope. We say that control must be established simultaneously in all stages, beginning with the mines and ending with commercial production of atomic fuel. We are told that the U.S.S.R. plan is merely a beginning, but we do not think that *any* plan can be a finished structure; as our experience grows there will be more and newer elements added.

"We see no reason to put an end to our quest for a solution of the question of atomic energy control. We, therefore, now propose that the General Assembly recommend to the Security Council that the Atomic Energy Commission continue its activity in the direction defined by the original instructions, and to draft conventions [treaties] for the banning of atomic weapons and for establishment of effective international control over atomic energy, these to be signed and entered into force simultaneously."

This modified the long-standing demand of the Soviets that the weapons be prohibited and destroyed immediately, without waiting for an agreement as to future control.

Two days later, WARREN S. AUSTIN, U. S. delegate, replied; extracts from his remarks follow:

"Let us look at the U.S.S.R. proposal and see if we can find out what it means. Its first recommendation seems to mean that the Atomic Energy Commission is to carry on in the direction laid down. The ordinary use of that language implies an approval of the majority reports of the Commission, now before us, and to carry on in the same direction, not in another. However, there has been so much violent disagreement within the Atomic Energy Commission about the meaning of the resolutions originally establishing it that it is hard to believe that the U.S.S.R. is now willing to continue along the same road.

"The Soviet resolution also speaks of 'effective international control'. What does this mean? Their contribution to the debate seems to indicate that they are not ready for effective international control and that they still adhere to the idea of nationalism and veto.

"The Soviet has demanded that stockpiles of weapons be destroyed. Does this mean security? No. What is the atomic weapon? Is it the casing—which anyone can make in any machine shop? Nobody has advocated the destruction of the vital energy that constitutes the menace—that is, the nuclear fuel that is inside it. Everybody has agreed that that has to be preserved for benevolent uses.

"There is another great field of work ahead for the Atomic Energy Commission. This is in the management of isotope separation plants, nuclear reactors, chemical extraction plants, and the preparation, storage and shipment of high-grade nuclear fuels."

On Oct. 20, the Political and Security Committee rejected the above-mentioned Soviet proposal (which would continue the Commission in existence but bypass the three reports already approved by its majority) and adopted a Canadian resolution which approved these reports as "constituting the necessary basis for establishing an effective system of international control of atomic energy to insure its use only for peaceful purposes and for the elimination from national armaments of atomic weapons". The Canadian resolution requested "the permanent members of the Atomic Energy Commission [Brazil, Canada, China, Russia, United Kingdom, United States] to consult in order to determine if there exists a basis for agreement on international control", meanwhile calling upon the Commission "to resume its sessions, to survey its program of work, and to proceed to further study of such of the subjects remaining as it considers to be practicable and useful".

On Nov. 4, the United Nations General Assembly approved this action of its Political and Security Committee, by a vote of 40 to 6. The Western Powers promised full support. As for trying to find a basis of agreement in private talks, Mr. VISHINSKY said, "We can immediately give a reply—there is no such basis."

Low-Temperature Properties of Lead-Base Solders and Soldered Joints

By R. I. Jaffee

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SOFT-SOLDERED ASSEMBLIES are used for containers or conduits that hold refrigerants or liquefied gases and for other articles subjected to subzero temperature. To obtain information about the solders that are best suited for such applications, the National Lead Co. sponsored research at Battelle Memorial Institute concerning the mechanical properties of some of the common lead-base solders at low temperature. The results are reported here.

It has been known for almost 25 years that pure tin is brittle at low temperatures. Lead, on the other hand, has good ductility at temperatures as low as -420°F . Until recently, nothing was known about the properties of tin-lead alloys and other common solders at subzero temperature.* Apparently, no information at all has been published concerning the low-temperature strength of soldered joints.

Solders Tested—The following alloys were tested for tensile strength and elongation at 75, -20 , -75 , and -295°F : 2.5% Ag, 97.5% Pb; 5% Sn, 95% Pb; 50% Sn, 50% Pb; and 1.6% Ag, 15% Sn, 83.4% Pb.

Charpy impact tests were made at the same temperatures on the same solders and, in addition, on 10% Sn, 90% Pb and 15% Sn, 85% Pb solders.

Soldered joints were tested at the same temperatures on sections of copper tubing soldered together with 2.5% Ag, 97.5% Pb; 5% Sn, 95% Pb; 25% Sn, 75% Pb; and 50% Sn, 50% Pb solders.

*EDITOR'S NOTE: Two recent papers abstracted on p. 884 of this issue are the only previous sources of information about the mechanical properties of solders at subzero temperature.

Solders containing more than about 90% lead retain their ductility and increase in impact strength at temperatures as low as -295°F . In solders with as much as 50% tin, serious embrittlement and decrease in impact strength occur at subzero temperature. Breaking loads of soldered copper tubing at low temperatures are nearly independent of the kind of lead-base solder used for joining.

Specimens—Specimens for tensile tests on solder alloys were cast in permanent molds and had shouldered ends and a 2-in. gage section 0.5 in. in diameter. The Charpy impact specimens were cast in permanent molds in the form of bars containing five specimens. They were of the simple-beam variety with standard dimensions and with V-notches cast in place, as described by B. W. Gonser and C. M. Heath (*Transactions of the American Institute of Mining and Metallurgical Engineers*, Vol. 122, 1936, p. 349 to 371). Soldered-joint specimens were made from two sections of 1-in. I.D., $\frac{1}{16}$ -in. wall, hard copper tubing, soldered together using standard sleeve fittings.

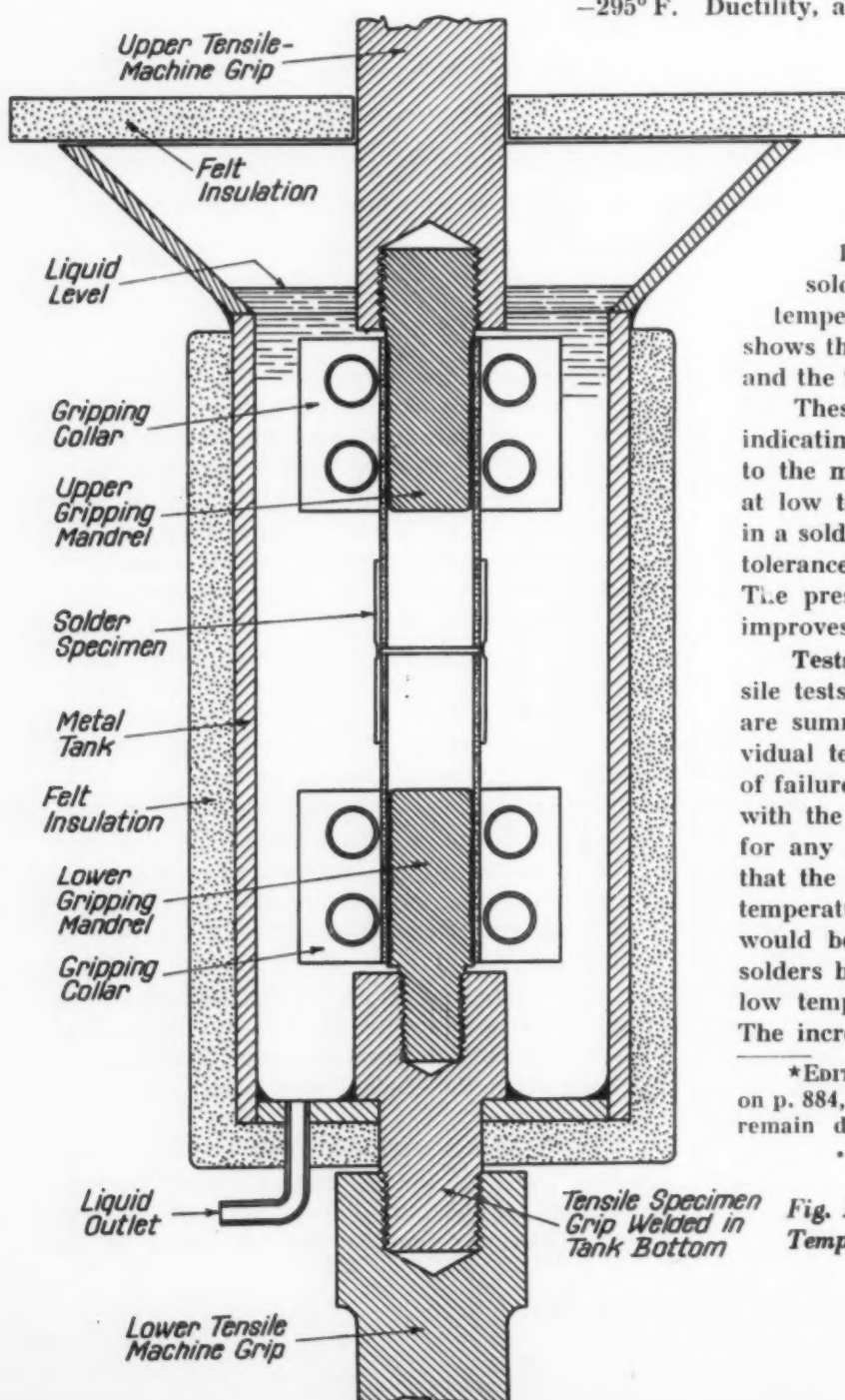
Testing Methods—All tests at subzero temperatures were made in liquid baths. At -20 and -75°F , a mixture of dry ice and acetone was used; at -295°F , boiling oxygen was the coolant.

Impact specimens were held at the desired temperature for at least 15 min. before testing. They were removed to an Amsler impact testing machine that was set for 58 ft-lb., and were then tested. The time interval between removing the specimen from the coolant and fracturing it was less than 2 sec.

Tensile tests were made in an insulated tank; the lower grip of the machine extended through the bottom of the tank. Grips were of the type recommended for shoulder-end specimens by the

American Society for Testing Materials. The rate of loading for the tension tests was 0.06 in. per min.

For tensile testing the soldered joints, the setup was somewhat more complicated, as shown in Fig. 1. A buttress-threaded mandrel was inserted in the tubing, and a collar was clamped over the tube, thus providing means for gripping the specimen. It was necessary to keep both the specimen and the clamp totally immersed in the coolant in order to prevent failure of the tubing at a level above that of the liquid.



For any one set of test conditions (that is, alloy and temperature), three tensile tests on solders, four tensile tests on soldered joints, and ten impact tests on solders were made.

Results of Tensile and Impact Tests

The results of tensile and impact tests on the solder alloys are shown in Fig. 2, where the data are arranged from left to right in the order of decreasing lead content in the alloys. Increasing the tin content (which adds a tetragonal phase to the structure) increases the brittleness of the alloys. The impact values of the alloys containing more than 95% lead increase with decreasing temperature. With lead contents from 85 to 90%, the impact values do not change much between room temperature and -295°F . When the lead content is as low as 50%, the impact values decrease to dangerously low values at -295°F . Ductility, as measured by the percentage elongation

in 2 in., follows the same general course as impact. The elongation of the 50% lead alloy decreases from the high value of 109% when tested at room temperature to 6% when tested at -295°F ., while the Charpy impact value decreases from

$15\frac{1}{2}$ to 1 ft.-lb. Tensile strength of the solder alloys increases almost linearly as the temperature decreases. The 50% lead alloy shows the greatest increase in strength at -295°F . and the 97.5% lead alloy shows the least increase.

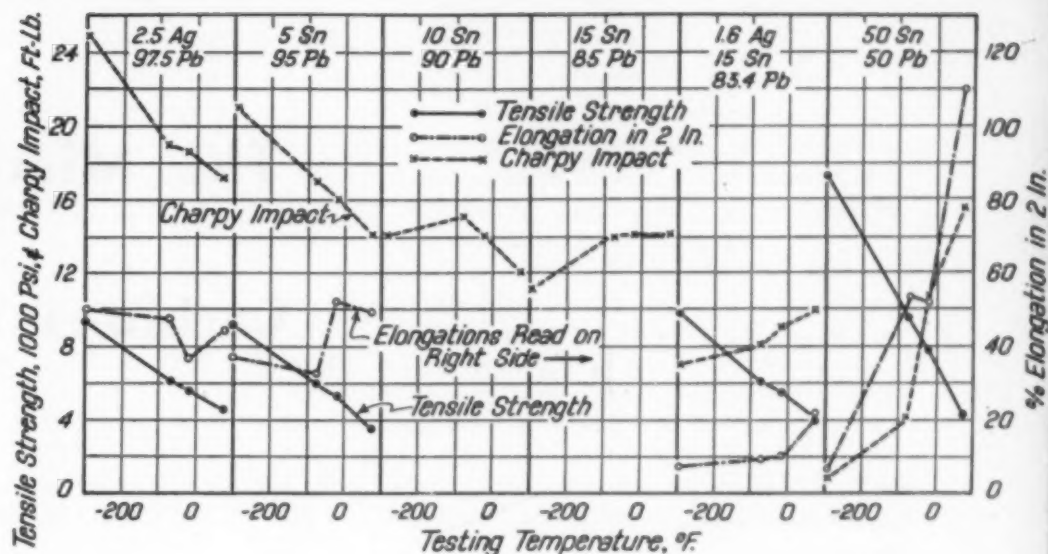
These results should not be interpreted as indicating that tin in any percentage is detrimental to the mechanical properties of lead-base solders at low temperatures. Only large amounts of tin in a solder alloy cause serious embrittlement. The tolerance for tin may be taken as 15%, at least.* The presence of tin, even in small percentages, improves the soldering properties greatly.

Tests on Soldered Joints—Results of the tensile tests on soldered tubing at low temperatures are summarized in Table I. The results of individual tests are given in order to show the type of failure in each test. In general, the joints made with the various solders broke at the same loads for any given temperature. This does not mean that the joints are all equally satisfactory at low temperatures. The value of the breaking load would be expected to reflect the strength of the solders but not their ductility, which decreases at low temperatures when the lead content is low. The increase in the breaking load of the joint as

*EDITOR'S NOTE: According to the paper abstracted on p. 884, alloys with more than 70 to 80% lead should remain ductile at temperatures near absolute zero.

Fig. 1 — Equipment Used for Determining Low-Temperature Tensile Properties of Soldered Tubing

Fig. 2 — Tensile Strength, Elongation and Charpy Impact of Lead-Base Solders From +75 to -295° F.



the temperature decreases is linear; the increase from room temperature to -295° F. amounts to about 30%.

The type of fracture of these soldered joints appears significant. The 50% Sn, 50% Pb soldered joints were stronger than either the tube or the fitting at temperatures down to -75° F. At -295° F. all the specimens of this composition failed in the soldered joint, probably because of low-temperature embrittlement. The solder with next higher tin content (25% Sn, 75% Pb) also failed at the soldered joint at -295° F. The 2.5% Ag, 97.5% Pb joints, which were somewhat weaker at room temperature than the other joints, failed in the soldered joints at temperatures down to -75° F., but were stronger than either the tubes or fittings (three specimens out of four) at -295° F., as were joints soldered with 5% Sn, 95% Pb alloy, tested at -295° F.

Conclusions

Soft solders that contain a high percentage of face-centered cubic lead retain their ductility and increase in impact strength at low temperatures. This is in agreement with recent concepts concerning the relation of structure to low-temperature properties. When the percentage of tetragonal tin becomes as high as 50%, serious embrittlement and decrease in impact strength occur. Tin contents up to 15% have no serious embrittling effect.

The increase in the tensile strength of solder alloys and in the breaking load of soldered joints

is linear with decreasing temperature. The solder containing the most tin (50%) shows the greatest increase in strength, and the solder with the most lead (97.5%) shows the least increase in strength. Breaking loads of soldered copper tubing at low temperatures are nearly independent of the kind of lead-base solder used. The impact strength and ductility of such joints would probably be influenced by low temperatures, in view of the properties of the solders alone.

Table I — Effect of Temperature on the Breaking Strength of Soldered Copper Tubes (1-In. I.D., 1/8-In. Wall)

50% SN, 50% PB		2.5% AG, 97.5% PB		5% SN, 95% PB		25% SN, 75% PB	
BREAKING LOAD, LB.	TYPE OF FAILURE	BREAKING LOAD, LB.	TYPE OF FAILURE	BREAKING LOAD, LB.	TYPE OF FAILURE	BREAKING LOAD, LB.	TYPE OF FAILURE
Specimens Tested at +75° F.							
9400	Tube	8600	Joint	9100	Joint	9400	Tube
9500	Tube	8400	Joint	9400	Joint	9500	Joint
9600	Tube	8600	Joint	8600	Fitting	9400	Tube
9400	Tube	8400	Joint	8900	Fitting	9400	Joint
Avg. 9500		8500		9000		9500	
Specimens Tested at -20° F.							
10,200	Tube	9600	Joint	9800	Joint	10,200	Tube
10,100	Tube	9800	Joint	9700	Tube	10,000	Joint
10,200	Tube	9700	Fitting	9900	Tube	10,000	Tube
10,200	Tube	9800	Joint	9400	Fitting	10,100	Tube
Avg. 10,200		9700		9700		10,100	
Specimens Tested at -75° F.							
10,500	Tube	10,400	Joint	9,300	Fitting	10,600	Tube
10,700	Tube	9,500	Joint	10,400	Tube	10,500	Tube
10,300	Tube	10,100	Joint	10,400	Tube	10,600	Tube
10,600	Tube	10,500	Joint	10,400	Tube	10,500	Tube
Avg. 10,500		10,100		10,400	(Tube)	10,500	
Specimens Tested at -295° F.							
12,500	Joint	12,800	Tube	12,500	Fitting	12,500	Joint
12,600	Joint	12,300	Fitting	12,500	Fitting	12,500	Joint
12,500	Joint	12,500	Tube	11,900	Fitting	12,600	Joint
12,500	Joint	12,700	Joint	12,600	Joint	12,600	Joint
Avg. 12,500		12,600		12,400		12,600	

The Microhardness of Carbides in Toolsteels

By L. P. Tarasov
Research Laboratories
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IN ORDER TO clarify the marked resistance to grinding of certain highly alloyed types of tool-steel, the Knoop hardness of the carbide particles was measured in three steels: plain carbon tool-steel containing 1.3% C; high-carbon high-chromium steel with a nominal composition of 2.2% C, 12.5% Cr and 1.0% V; and high-carbon, high-vanadium high speed steel with a nominal composition of 1.5% C, 12.0 to 14.0% W, 5.0% Cr, 5.0% V and 5.0% Co. The plain carbon steel was annealed in order to make the cementite massive enough to be indented; the other two steels were hardened and tempered.

The hardness was also determined for aluminum oxide, in the form of Alundum grains, in order to relate the hardness of the carbide particles to that of the abrasive used in grinding the steels. Indentations were also made in a hardened piece of oil-hardening toolsteel, to establish a comparison between the Knoop numbers and the Rockwell C-scale. The carbide particles in this steel were

so small and so uniformly distributed that the Knoop indentations represent an average hardness for both the carbides and the martensitic matrix. This corresponds to the condition existing when a standard Rockwell reading is obtained.

The tests were made with a Tukon tester, using a 25-g. load, in order to keep the indentations appreciably smaller than the carbide particles. With a 25-g. load, the indentations in the carbide particles were between 10 and 20 microns long. In most specimens, only one indentation could be made in a carbide particle. Because of the strong directional effects in aluminum oxide abrasive, a greater number of indentations were made in this material—generally not more than two or three per grain, however—so as to average out the orientation effects. The indentations were measured with filar micrometer and a Bausch & Lomb Research Metallograph equipped with oil-immersion apochromatic 80× objective. The Knoop values for these conditions are given in Table I.

To avoid possible confusion with other published hardness numbers, especially for aluminum oxide, it should be noted that the values quoted

in this table are for a 25-g. load, as indicated by the subscript in the hardness symbol K_{25} . Had the carbide particles been large enough so that a greater load could have been used for testing, the resulting hardness numbers would have been appreciably lower, and the percentage decrease would have been greatest for the hardest substances. For instance, the hardness of aluminum oxide, which is 2440 for a 25-g. load, is 2000 for a 100-g. load. It should be noted also that the hardness number

Table I—Microhardness of Abrasive Grains and of Carbides in Steel

MATERIAL INDENTED	KNOOP HARDNESS (K_{25})		NO. OF INDENTS
	AVERAGE	RANGE	
Oil hardening toolsteel (R_C 60½)	790	770-800	4
Cementite (in plain carbon toolsteel)	1150	1060-1240	8
Complex chromium-iron carbides (in high-C high-Cr steel)	1820	1690-1960	7
Aluminum oxide (in regular Alundum abrasive)	2440	1900-2920	19
Complex vanadium carbides (in high-C, high-V high speed steel)	2520	2340-2760	8
Silicon carbide (calculated from data for 100-g. load)	3590	3070-3980	200

is influenced considerably by the objective used in measuring the length of the indentation when the indentation is extremely small. The variation of Knoop hardness number with load and conditions of observation has been discussed in detail elsewhere ("Determination of Knoop Hardness Numbers Independent of Load", by L. P. Tarasov and N. W. Thibault, *Transactions*, Vol. 38, 1947, p. 331 to 353).

The data in Table I show clearly that the complex chromium-iron carbides in high-carbon high-chromium steel are about 50% harder than cementite. The complex vanadium carbides are more than twice as hard as cementite and are even slightly harder, on the average, than aluminum oxide. Because of the wide variation in the hardness of aluminum oxide abrasive with crystallographic orientation, it is considerably softer than the complex vanadium carbides in some

directions, but in other directions it is harder. This hardness relation probably accounts for the difficulty in grinding high-carbon, high-vanadium high speed steels with aluminum oxide.

It might be assumed that silicon carbide, which is appreciably harder than either aluminum oxide or the complex vanadium carbides, would be the logical abrasive to use in grinding high-carbon, high-vanadium high speed steels. However, extensive experimentation has shown that although such steels can be ground with silicon carbide wheels, more satisfactory results can be obtained by the proper choice of aluminum oxide wheels and of grinding conditions. The reason is that the advantage of grinding the extremely hard carbide particles with silicon carbide is more than offset by the disadvantage of grinding the martensitic matrix with this abrasive, for which purpose it is known from experience to be unsuitable. ●

Correspondence

Discovery of Aluminum Brass

CHISLEHURST, KENT, ENGLAND

To the Readers of METAL PROGRESS:

In connection with the article "Milestones in the Brass Industry", by D. R. Hull (*Metal Progress* for June) some information may be of interest concerning the origin of the aluminum brass now widely used for condenser tubes.

The discovery of aluminum brass was a revolutionary advance in alloys for condenser tubes and was one unforeseen byproduct of some work on the casting of strip brass ingots, a piece of "operational research" that I was carrying out in the 1920's at Woolwich for the British Nonferrous Metals Research Association. At that time, aluminum was a metal carefully avoided in the brass foundry (it still is, in some places — even in America) and the defects that it could cause were investigated. It was found, however, that when castings were being made by the Durville process, aluminum was beneficial and could give ingots of perfect surface quality. The next step was the sideline investigation of the aluminum brasses, and it was evident that even a small content of aluminum greatly reduced the rate of oxidation and corrosion of the alloys. In one experiment, a cube of

aluminum brass was heated to the melting point and assumed a spherical shape with projecting sharp corners without collapsing, entirely because of the strength of the surface film of aluminum oxide. A study of the ternary system showed that the fabricating properties of the then-common condenser tube alloy could be accurately reproduced in a brass containing 76% Cu, 22% Zn, with 2% Al, and this basic composition has not so far been improved. The subsequent research on this alloy perfected it for use in condenser tubes. That, in brief, is the story — a useful example of the unpredictable benefits that may arise out of metallurgical research.

The detailed results and the number of the British patent can be found in my paper "The Aluminum Brasses", which was published in the *Journal of the Institute of Metals*, V. 43, 1930, p. 163 to 180. The work is also summarized in the book "The Casting of Strip Brass Ingots", by Genders and Bailey, published by the British Nonferrous Research Association. For the initiative in developing the condenser tube application, most credit is due to Dr. R. S. Hutton, who was then the Director of the Research Association.

R. GENDERS
Consultant Metallurgist

High-Purity Helium for Welding*

AMARILLO, TEXAS

To the Readers of METAL PROGRESS:

The Bureau of Mines has recently announced that its entire production of helium is to be known as "welding-grade" helium and will have a purity of 99.8% or more.

For the past 18 years, the production of helium by the Bureau of Mines has been confined chiefly to gas having a purity of 98.2%, and this has been satisfactory for use in lighter-than-air craft and for the various medicinal uses. Also, helium of 98.2% purity was satisfactory when used in the shielded-arc welding of magnesium and stainless steel. However, it was not entirely satisfactory in the welding of aluminum.

When welding-grade helium is used in welding aluminum, one fifth to one third more helium is normally required than argon when argon is used. But by using helium, greater penetration is obtained and much faster welding is possible than with argon, so that on many jobs the over-all cost per inch of weld is favorable for the use of helium as a shielding medium.

Before conversion of its helium plant for the production of welding-grade helium, the Bureau of Mines made extensive surveys to ascertain the effects of impurities. These studies resulted from the desire to supply the best possible inert gas for welding purposes. It was found that, to make entirely satisfactory welds on aluminum, the helium should be as free as possible from hydrogen, nitrogen, water vapor and oil vapor.

Many fabricators using the shielded-arc process have the idea that if they do not obtain at once a satisfactory weld using the shielded-arc process, the difficulty can be overcome by increasing the flow of inert gas. Usually this is a mistaken idea. Increasing the gas flow above a certain maximum will not improve the condition of the weld, and it may even produce porous welds.

When welding-grade helium is used in the shielded-arc process, it can produce aluminum welds having a bright surface, excellent fusion and penetration are obtained, and the gas used is not excessive. As an example, the Bureau regularly demonstrates the production of corner welds on 1/8-in. aluminum using 11 to 12 liters of welding-grade helium per min. and welding at the rate of 25 in. per min. Comparable welds using argon require approximately 8 liters per min., but slower

*EDITOR'S NOTE: Additional information on the use of helium in inert-gas-shielded arc welding is given in the article on p. 833 of this issue. For a description of two new welding processes that use inert gas for shielding, see the abstracts on p. 886 to 890.

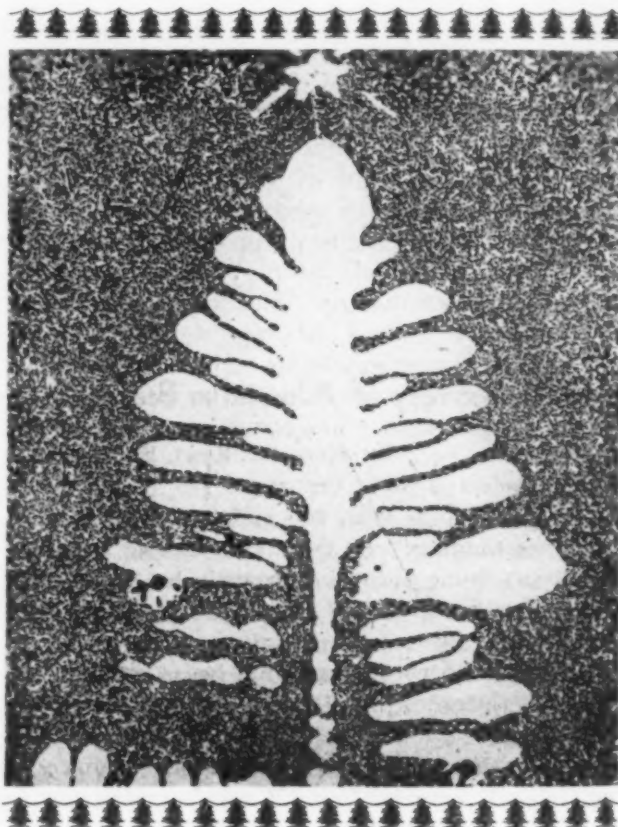
welding speeds are necessary to obtain good penetration.

Welding-grade helium is now obtainable through distributors of compressed gas in nearly every part of the country. The Bureau is confident that this helium will be satisfactory for most of the requirements of the inert-gas welding industry.

The Bureau of Mines plans to publish a series of articles on the use of helium in inert-gas arc welding, in which technical data showing tensile strength, micrographs and radiographs of welds will be included.

WILLIAM A. MAYS
Metallurgical Engineer
Bureau of Mines

Season's Greetings!



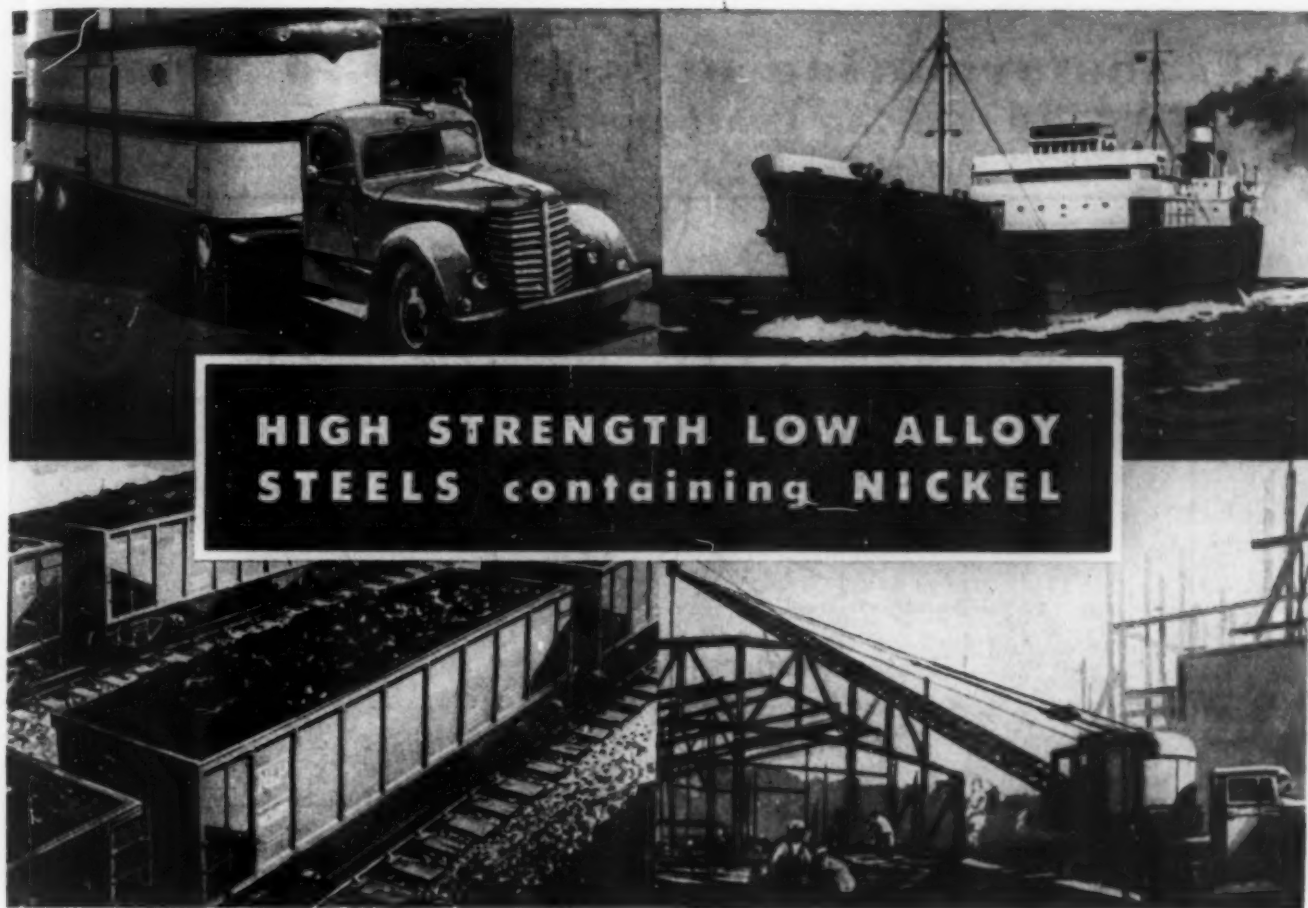
A Very Merry Christmas
From the Metallographic Department
Of the Permanente Metals Corporation

SPOKANE, WASH.

To the Readers of METAL PROGRESS:

The alpine fir tree from the great Northwest is a dendritic arm of aluminum in a background of aluminum-silicon eutectic, at 75X. The star, of course, was an appropriate afterthought.

L. J. BARKER
Metallographer
Permanente Metals Corp.



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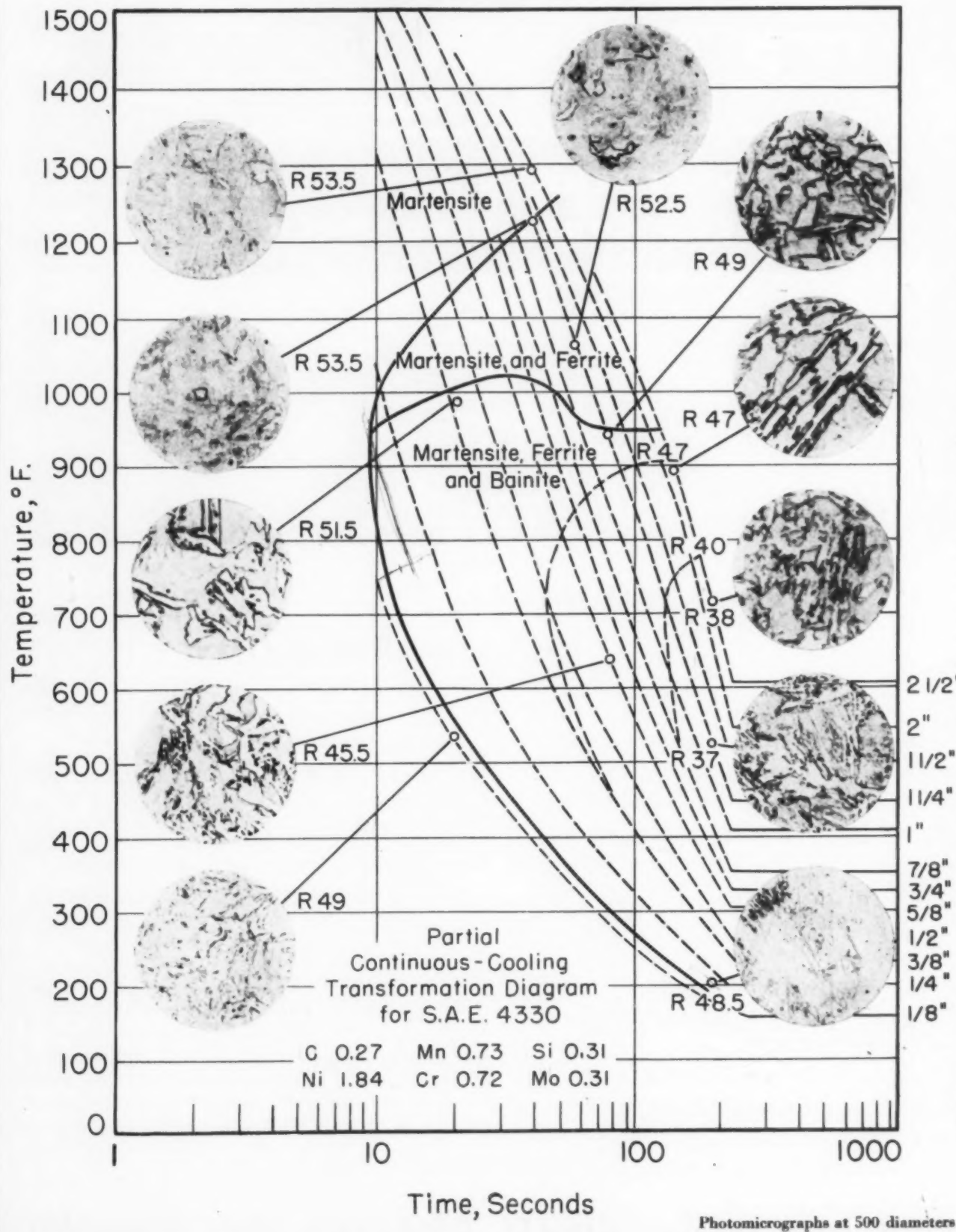
THE INTERNATIONAL NICKEL COMPANY, INC.

**67 WALL STREET
NEW YORK 5, N. Y.**

December, 1948; Page 848-A

Transformation of S.A.E. 4330 Steel During Continuous Cooling

By C. A. Liedholm and Associates, Curtiss-Wright Corp.



Structural Changes

During Continuous Cooling

By Carl A. Liedholm
Quality Manager
Piasecki Helicopter Corp.
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Information shown on continuous cooling diagrams and data obtained from flat Jominy bars can be used to advantage in planning practical heat treatments. This article tells how it is done. The author makes a plea for the extension of such data to include slower cooling rates than heretofore.

CONTINUOUS COOLING is obviously one of the most important processes in heat treating, for whenever a piece of steel is heated above its critical range, it must sooner or later be allowed to cool, and it is the continuous cooling that determines whether we harden, normalize, or anneal the steel.

A hardenability test such as the Jominy test shows the result of a large number of continuous cooling rates, from one that is very fast to one that is fairly slow. But it does not show the sequence of the structural changes, nor their timing.

Isothermal transformation diagrams show the sequence and timing of structural changes that occur in steel when it is quenched suddenly to any subcritical temperature and is held at that temperature. However, isothermal transformations differ from transformations during cooling, in some important respects. To base predictions of cooling transformations on isothermal diagrams, one must be familiar with the relationship between isothermal and cooling transformations. Much theoretical study has been devoted to this relationship during the last few years, but (to a practicing metallurgist) it is a tedious and often hazardous undertaking to arrive at a correct interpretation of isothermal data in terms of structural changes during continuous cooling. Diagrams showing the

actual temperatures and the actual timing of structural changes during continuous cooling are needed to round out the metallurgist's understanding of one of the most important processes that confront him in his daily experience.

It is my intention to show how simply such diagrams can be obtained by a well-equipped laboratory. I shall then discuss some uses for the present type of diagrams, show how the information they give can be supplemented with flat Jominy bars and, finally, indicate how one can develop them further to increase their usefulness and

to provide the metallurgical practitioner with a long-needed aid in his efforts to predict heat treating results and to plan heat treating cycles with the maximum prospect of attaining precisely the expected results.

For the benefit of those not entirely familiar with either the Jominy test or the S-curve, I shall begin with a brief description of these two methods for expressing the transformation characteristics of steel, and then show how the two methods can be combined to give continuous cooling transformation diagrams.

The Jominy end-quench hardenability test has been standardized by both the S.A.E. and A.S.T.M. for several sizes of specimen. Almost always a 1-in. diameter cylinder, 4 in. long, is used — normalized, reheated to the specified hardening temperature, and then hung vertically in a fixture, with a water column directed against the bottom face of the specimen. This face is referred to as the quenched end. After 10 min. end-quenching, two parallel, diametrically opposite flats are ground along the entire bar, whereupon Rockwell hardness tests are taken at $\frac{1}{16}$ -in. intervals over the first 1-in. length containing the quenched end, and at arbitrary, greater intervals along the rest of the flat. The hardness is then plotted against distance from

quenched end, and the resulting curve gives a measure of the hardenability of the bar.

The cooling rates at different distances from the quenched end have been determined precisely by Boegehold and Weinman for 1-in. round specimens, and have been correlated by the same authors with cooling rates for different radial positions of water quenched and oil quenched bars and plates on the basis of the cooling rate at 1300° F. These data permit the prediction of the hardness distribution across a transverse section of a quenched bar or plate from a Jominy hardness curve. (See *Metal Progress* data sheets, January and February 1947.) Figure 1 shows Jominy hardness curves for two heats of S.A.E. 6130 steel, as quenched and after tempering at 1200° F. for 2 hr.

An isothermal transformation diagram (or S-curve or TTT-curve) is produced in the following way: A series of small and thin pieces of steel are quenched into a salt or lead bath held at constant temperature. At suitable intervals, one piece at a time is withdrawn from the bath, quenched to room temperature, and examined under the microscope. The untransformed austenite will have changed to martensite by the final quench, whereas stable structures formed as a result of the constant-temperature transformation will remain unaffected by that quench. By repeating this process, using a series of baths at different constant temperatures, enough data can be obtained to plot the beginning and end of the different types of microstructural changes that occur at each isothermal transformation temperature. The salient point of the technique is to stop a transformation at different points of its

progress by quenching the specimen to room temperature, and to plot the time when transformation first was observed at each isothermal temperature, and the time when the structural change had ceased.

Figure 2 shows an isothermal transformation diagram for S.A.E. X4130 steel.

Analysis of Jominy Bars

The *continuous* cooling transformation diagram is obtained from Jominy bars in four steps.

First, one obtains accurate cooling curves for a number of distances from the quenched end of the bar, and plots the curves on large graph paper.

Second, one end-quenches a number of Jominy specimens for various periods of time, terminating each end-quenching period by complete immersion quenching of each specimen.

Third, one examines the microstructure at different distances along these Jominy bars, observing the distance from the quenched end and the time of end-quenching required for the appearance of characteristic structural changes (such as ferrite, bainite, or pearlite formation).

Fourth, to locate the temperature of structural change at any distance from the quenched end, one finds the intersection between the cooling curve for that distance and the ordinate through the end-quenching time which brought about the change. All that remains is to connect all points on the graph which signify the same change, with a faired line.

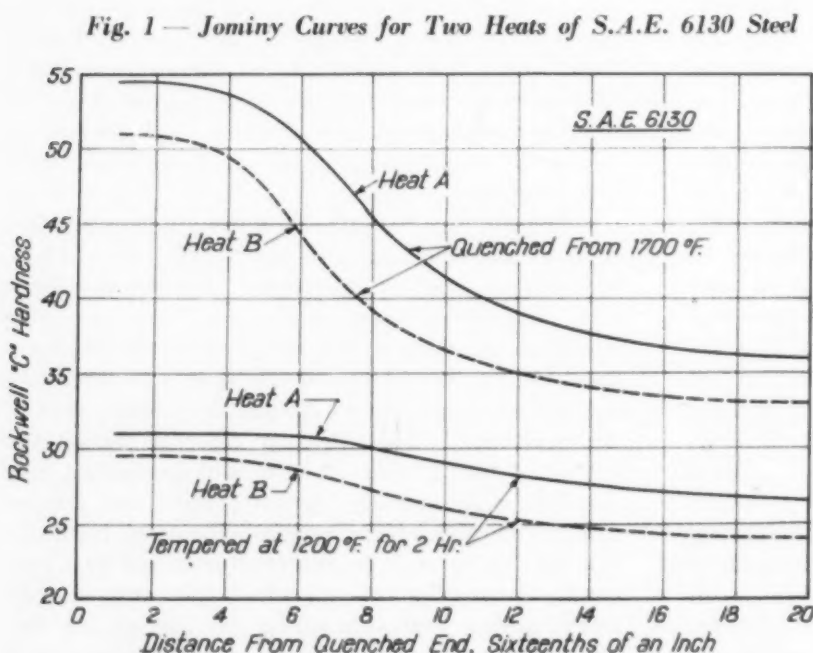
The salient point of this technique is similar to the S-curve technique, in that transformations are stopped or "frozen" through quenching at various stages of progress, and the resulting structures are then studied under the microscope. Thereupon,

the time when each structural change first was observed is plotted in the same way as isothermal structural changes, the only difference being that the cooling changes are shown associated with a series of cooling curves.

A continuous cooling transformation diagram for S.A.E. 4330 steel is printed on p. 848-B of this issue. Because it is very important to the following discussion that this figure be fully understood, I shall present a rather detailed explanation of it.

Time is plotted horizontally, temperature vertically. The slowest cooling curve is for the 2½-in. distance, and the fastest cooling curve is for the ⅛-in. distance.

To visualize the meaning of the graph, follow, for instance, the vertical line that intersects the



30-sec. distance on the horizontal axis. It intersects the cooling curve for $\frac{1}{8}$ in. from the quenched end at 450° F., $\frac{1}{4}$ in. at 575° F., $\frac{3}{8}$ in. from the quenched end at 785° F., and so on. Those are the temperatures of these distances from the quenched end of the Jominy bar after 30 sec. end quenching.

The upper solid line in the diagram for S.A.E. 4330 is the ferrite beginning line, since it shows that metallographic examination of a specimen quenched by full immersion, for instance after 30 sec. end quenching, disclosed rejected ferrite only at those distances from the quenched end whose cooling curves after 30 sec. had crossed the upper solid line.

Hence, after 30 sec. end quenching, rejected ferrite existed only between the $\frac{1}{4}$ and $\frac{3}{4}$ -in. distances from the quenched end, because the cooling curves corresponding to these positions on the Jominy bar had crossed the ferrite beginning line—but not at somewhat less or greater distances, because those cooling curves had not crossed the ferrite beginning line after 30 sec. end quenching.

Applying the same reasoning, the lower solid line is the bainite beginning line, since it shows whether or not bainite had formed at the various distances from the quenched end after various end-quenching periods. For instance, after 25 sec. end-quenching, bainite had formed at the $\frac{1}{4}$, $\frac{3}{8}$, and $\frac{1}{2}$ -in. distances from the quenched end because their cooling curves had crossed the bainite beginning line after 25 sec., but bainite had not formed at any other distance from the quenched end because none of the other cooling curves had crossed the bainite beginning line after 25 sec. The hardness contours connect points of equal hardness. The diagram can be described as a partial time-temperature-transformation diagram for continuous cooling.

Practical Applications

Now let us see what type of problems of metallurgical and engineering importance can be solved with the aid of the present diagrams.

First, suppose we had performed a post-mortem analysis of a structural member that showed clear evidence of having failed in fatigue. Assume that the member was made of S.A.E. 4330 steel on which the required data are available, and

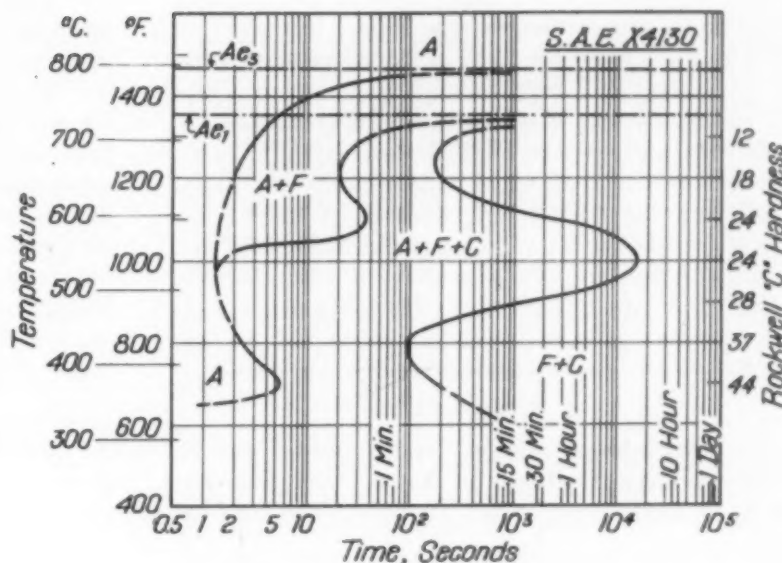


Fig. 2 — Isothermal Transformation Diagram for S.A.E. X4130 Steel (0.33% C, 0.53% Mn, 0.90% Cr, 0.18% Mo). Steel austenitized at 1550° F. Grain size, 7 to 8. (Atlas of Isothermal Transformation Diagrams, U. S. Steel Corp.)

that the microstructure contained an estimated 20% rejected ferrite, and that a competent metallographic analysis showed no other defective metallurgical condition. The rest of the structure will be assumed to have been tempered martensite, the specified heat treatment being to quench to a fully martensitic structure and temper at 1120° F. Our problem will be to determine if the ferrite was likely to have been responsible for the failure.

We are not willing to point an accusing finger at the ferrite without first putting it to a test. So we have conducted the following experiments: Air cooling curves have been obtained for the midpoint of the R. R. Moore specimens which we use to estimate the basic fatigue strength of the steel. The average air cooling curve has been superimposed on the cooling transformation diagram of the same steel and several series of specimens have been air cooled before quenching for different periods, estimated with the aid of the diagram so as to give various amounts of free ferrite. We have then plotted fatigue strength against air cooling period. We also performed the same experiments with tensile and impact specimens.

Each set of specimens of the same kind and quench delay was air cooled in a symmetrically loaded holder before the holder and specimen were quenched together. The cooling cycle of the $\frac{1}{4}$ -in. diameter shoulder type of tensile specimen used in this particular investigation was approximated by the $\frac{1}{8}$ -in. Jominy distance, and that of the R. R. Moore fatigue specimens by the $\frac{2}{3}$ -in.

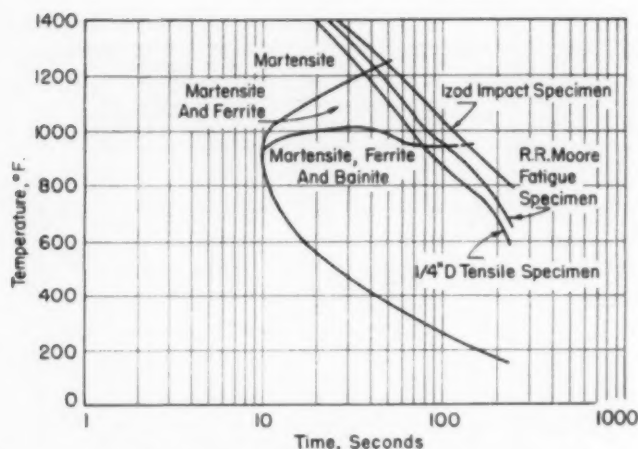


Fig. 3 — Air Cooling Curves for Tensile, Impact, and R. R. Moore Fatigue Specimens of S.A.E. 4330 Superimposed on the Continuous Cooling Transformation Diagram

distance. The Izod specimens cooled more slowly than the 2½-in. position on the Jominy test bar.

Figure 3 shows the cooling curves superimposed on the cooling transformation diagram; Fig. 4 shows endurance limit plotted against the duration of air cooling and, also, typical microstructures. Figure 5 shows percentage of ferrite in ¼-in. tensile specimens plotted against quench delay or air cooling time, and allows Fig. 4 to be interpreted approximately in percentage of ferrite versus fatigue strength. It shows that 20% ferrite corresponds to about 120 sec. air cooling, and that, therefore, the fatigue strength was not lowered by more than a few per cent — if at all.

Considering that a 60° notch with a radius of 0.002 in. will decrease the fatigue strength of a Moore specimen by about 75%, and that most structural members contain shoulders or recesses or other types of stress raisers, it is obvious that the 20% ferrite which our investigation had disclosed in the microstructure should hardly be included among even the secondary causes of failure. However, we were discussing a steel of high hardenability that also is fairly high in ferrite-forming alloying elements; the conclusions we arrived at might have been different for a different type of steel. Similar analysis concerning various types of steel would help to clarify many problems of this sort, particularly if the analysis included a consideration of the related ease or difficulty of welding the steels.

Figure 6 shows how ferrite affects the tensile strength and hardness of the same steel in the as-quenched and as-tempered conditions.

The example just cited concerned ferrite that formed because the quench had been delayed excessively. Ferrite may form in this way when a hot part has to be loaded into a cold fixture or die before quenching, or when the distance between the furnace and quenching bath is too

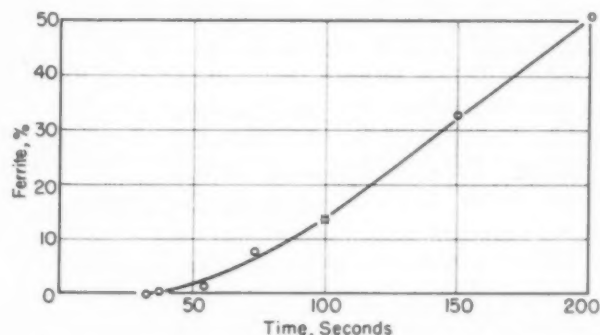


Fig. 5 — Percentage of Ferrite in ¼-In. Tensile Specimens Versus Pre-Quench Air Cooling Time

great, or the handling equipment is cumbersome and slow. If the steel is one that reacts rapidly and if the rate of air cooling is just less than the critical quenching speed, the same phenomenon of ferrite formation may be observed.

For ordinary quenching conditions, however, a situation like that described is not likely to occur. Consider, for instance, the continuous cooling transformation behavior of a steel such as S.A.E. 6115. This steel does not have high hardenability; yet, when quenched from 1700° F., it cannot react in less than about 8 sec. — and then

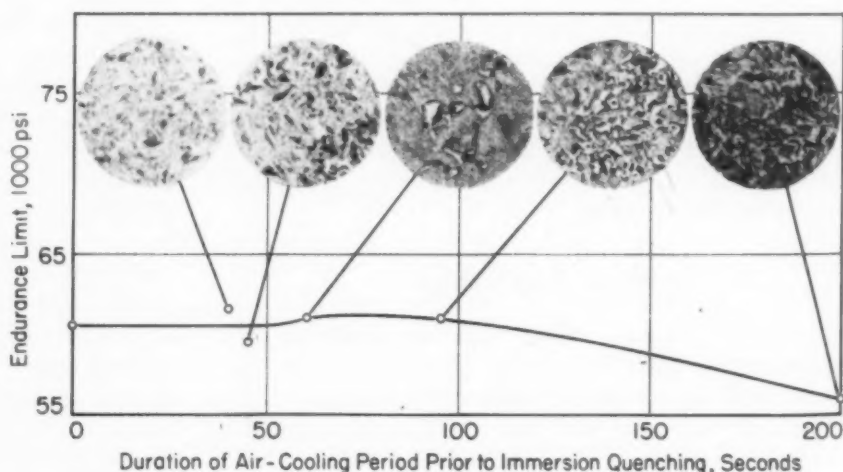


Fig. 4 — Endurance Limit of S.A.E. 4330 Steel Plotted Against the Duration of the Air Cooling Period Prior to Immersion Quenching

only if it has reached temperatures between 950 and 1150° F. If the cooling cycle is faster, the result will be full quenching, and if it is slower, transformation will not have begun after that length of time.

This kind of information is important in connection with the diagnosis of unsatisfactory quenching results. Most metallurgists probably have been asked by heat treaters at some time or another how quickly they have to immerse a part in the quenching bath to avoid transformation during the short air-cooling period preceding the quench. Usually, the correct answer is that you

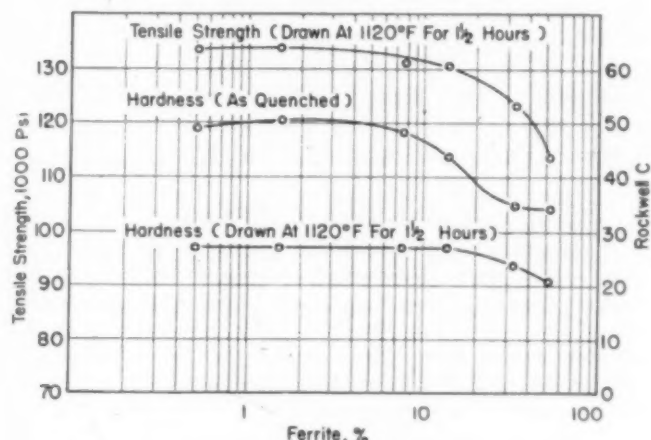


Fig. 6 — Effect of Pre-Quench Ferrite on the Hardness and Tensile Strength of S.A.E. 4330 Steel

don't need to hurry as much as you might think. However, it is obvious from the preceding discussion that thin sections of steel of low hardenability can react during transfer from a furnace to a quenching bath.

There are times when manufacturing procedures render it exceedingly difficult to quench steel to a fully martensitic structure. Consider, for instance, the heat treatment of large welded assemblies. Since steel becomes increasingly susceptible to welding cracks as its carbon content increases, it is customary to select steel with a maximum carbon content of about 0.35% for welding manufacturing. For this and other reasons, S.A.E. 4130 is commonly used in aircraft weldments. Welding rods usually are made from steel with about 0.10 to 0.15% max. carbon. The welds made from such rod in S.A.E. 4130 steel never have high hardenability.

Usually weldments must either be oil quenched or merely normalized. Most weldments contain sections that vary considerably in thickness. If such a weldment is quenched in oil, part

of it may be slack-quenched while other parts will be fully quenched. The heat treatment of such structures to uniform mechanical properties presents complicated metallurgical questions. To attain control of the results, the metallurgist must be able to estimate the slowest cooling rate that results from the quenching. This information he can obtain by finding the position on a Jominy bar whose hardness matches the hardness of the slowest cooling section of the welded structure, in accordance with Boegehold's method. The Jominy bar can be tempered at any desired temperature to obtain an estimate of the hardness variations that would exist in the weldment after heat treating.

Let us assume, however, that it will be necessary not only to know the hardness, but also to obtain an estimate of the tensile properties. A rather obvious method for obtaining this information consists of preparing a flat end-quench bar from which tensile specimens can be machined after it has been end-quenched and drawn. This method has been developed and used by the writer in connection with studies of heat treating propeller blades at Curtiss-Wright. Figure 7 shows the design of a flat specimen; Fig. 8, the end-quenching operation.

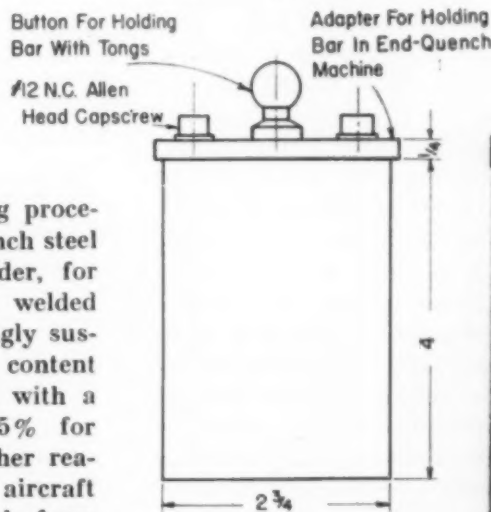
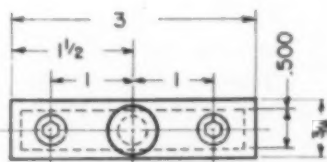


Fig. 7 (above) — Design and Dimensions of the Flat Jominy Bar, From Which Tensile Specimens Can Be Machined After the Bar Has Been End-Quenched and Tempered

Fig. 8 (below) — End Quenching the Flat Jominy Specimen of Fig. 7. Water is directed upward by a special orifice.



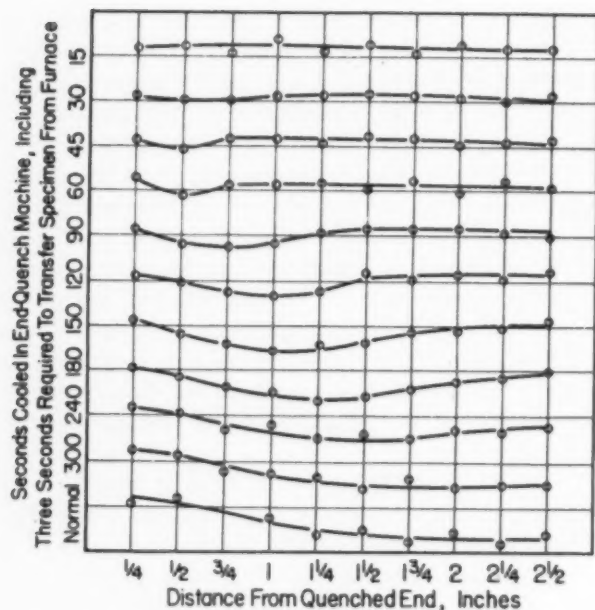


Fig. 9 — Effect of Pre-Quench Air Cooling Time on the Tensile Strength of S.A.E. 4330 Steel

Figure 9 shows tensile strength results obtained from a flat Jominy bar that had been end-quenched for various periods of time and subsequently tempered. In the figure, the height of a block represents 20,000 psi., and the abscissa that intersects each curve of tensile strength versus Jominy distance is at 130,000 psi. Thus, the effect, on the tensile strength of S.A.E. 4330 steel, of widely varying cooling cycles and air cooling periods preceding the quenching operations can be evaluated. The effects of the same variables on other mechanical properties have been evaluated also.

Careful cooling curve determinations have been made for the flat end-quench bar, and for the 1/2-in. diameter, 4-in. long end-quench bar that was used for the continuous cooling transformation diagrams, as well as the standard 1-in. Jominy bars. Many determinations were made for some positions on the 1/2-in. diameter end-quench specimen between 1943 and 1946. When Weinman and Boegehold published their cooling curves for standard Jominy bars, it obviously became a matter of great interest to compare their data with the curves we had obtained.

Figure 10 shows a reproduction of Boegehold and Weinman's curve for cooling velocities at 1300° F.; the writer's data is represented by the solid black dots. Keeping in mind that only single determinations had been made for each plotted point of the 1-in. Jominy bar, the agreement is excellent. Since as many as 20 cooling curves were obtained for some positions of the 1/2-in.

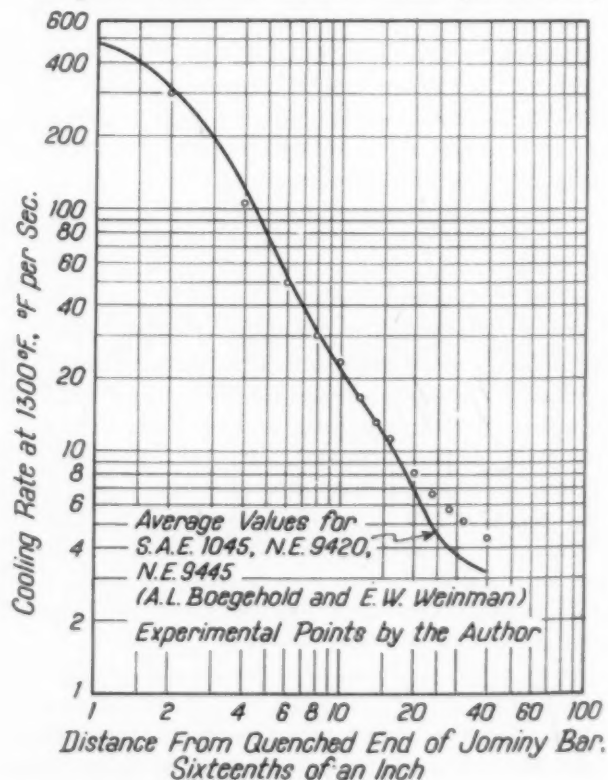
diameter Jominy bar, the data on that bar are probably even more reliable.*

Thus far, we have concerned ourselves with the time required for structural changes to begin during cooling and with the effects of partial transformation on the properties of the steel. It has been shown that such phenomena can be studied with the aid of the cooling transformation diagrams supplemented by flat Jominy bars. These problems are all associated with the beginning and early stages of transformation or, in other words, with the left-hand part of the isothermal or continuous-cooling diagrams.

It has been shown that the actual events can be estimated more accurately from the continuous cooling diagrams than from the S-curves. In justification of this statement, Fig. 11 has been included. This figure is a reproduction of the U. S. Steel Corp.'s isothermal diagram for S.A.E. 4130 steel, with the beginning-lines for

*Our cooling velocities obtained for the slowest cooling part of the Jominy bar are higher than shown by Boegehold and Weinman, probably because our data were obtained on steel of much higher hardenability than that used by Boegehold and Weinman. Thus, the specimens used by them transformed to a greater extent than ours toward the air-cooled end, with the result that their cooling curves were retarded by the release of the latent heat of transformation.

Fig. 10 — Boegehold and Weinman's Cooling Rates Compared With the Author's Data (Black Dots)



transformation during normal continuous cooling starting at 1700° F. traced in as a medium dotted line. The differences in positions of these lines cannot be explained solely on the basis of different austenitizing temperatures.

The End of Transformation

Having dwelt on the practical significance of the transformation beginning lines, we should consider, at least briefly, the significance of the transformation ending lines that are shown on isothermal diagrams but as yet have not been determined for continuous cooling. These lines are of as much importance to the annealing of steel or the postheating of welds as the beginning lines are to the hardening of steel.

To anneal steel, it is customary to heat it either above or slightly below A_{c3} , and then to cool it below A_{c1} , holding it in this "pearlite region" until transformation is complete. The time of holding and the accuracy with which the temperature is maintained are probably the most important factors in the majority of annealing cycles. Fortunately, both the time and temperature of transformation in the pearlite region are, for practical purposes, independent of the manner in which the steel is allowed to cool from the austenitizing temperature to the transformation temperature. Both time and temperature, therefore, can be determined with sufficient accuracy from isothermal diagrams.

The postheating of weldments is far too complicated, and also too controversial, a metallurgical problem to be entered into, except in a very superficial manner. I should like to mention, however, that some leading authorities have recommended that weldments be postheated in a manner such as to effect complete transformation to bainite in order to prevent the formation of martensite during cooling to room temperature. It has been suggested that the minimum time and the optimum temperature for bainite transformation be determined from the isothermal transformation diagrams of the steel being welded.

Since it has been thoroughly demonstrated that under-bead weld cracks practically always develop in martensitic areas, but since weldments almost never can be transformed in a strictly isothermal manner, it is important to know

whether or not transformations at constant temperatures in the bainite region are affected by the preceding cooling cycles. This question has also been studied and it has been demonstrated conclusively that structural changes during continuous cooling have a profound effect on subsequent isothermal transformation in the bainite region.

Figure 12 gives an idea of this effect. It shows the end-of-transformation lines for S.A.E. 4340 and 4360 steels, reproduced from U. S. Steel Corp.'s "Atlas of Isothermal Transformation Diagrams".

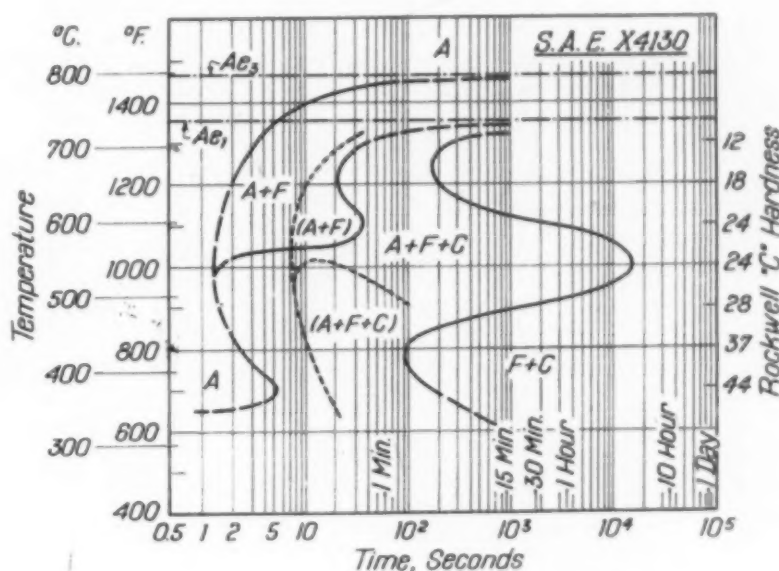


Fig. 11 — Isothermal Versus Continuous Cooling Transformation. The heavier lines denote isothermal reactions (U. S. Steel's Atlas). The lighter, dashed lines denote the beginning of transformations during continuous cooling. Lettering enclosed in parentheses refers to structures formed during continuous cooling. Data obtained on two similar heats of S.A.E. X4130 steel. Isothermal specimens: 0.33% C, 0.53% Mn, 0.90% Cr, 0.18% Mo, austenitized at 1550° F. Continuous cooling specimens, containing 0.31% C, 0.47% Mn, 0.92% Cr, 0.17% Mo, were austenitized at 1700° F.

Superimposed on this graph are micrographs of transformation structures obtained in a strictly isothermal manner, and also those obtained by constant temperature transformation preceded by slow cooling. It can be seen from this figure that the strictly isothermal transformation structures verify the lines of the diagram, but when the transformation has been preceded by continuous cooling, it takes more than 12 times longer for the structural changes to proceed to completion—exactly how long has never been determined.

Isothermal transformation diagrams had been known for more than 15 years before this difference between strictly isothermal transformation

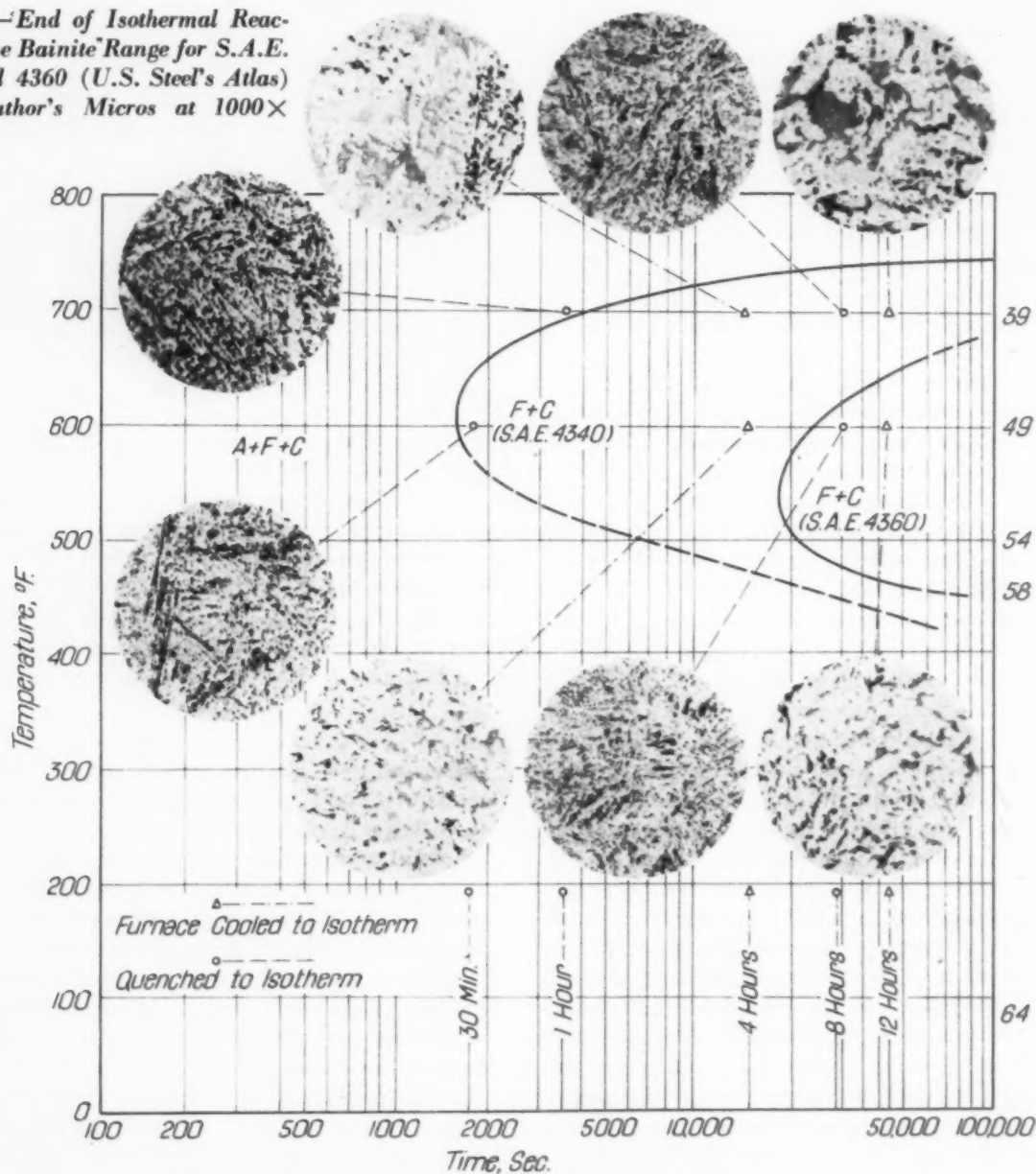
and constant-temperature transformation preceded by slow cooling was first demonstrated.

While engaged in this development work, I could not help envisioning and hoping for the future expansion of the continuous cooling transformation diagrams to include cooling of at least a week's duration. It would require the resources of a large organization to compile such data, but I believe that the reward in the form of practical applications would be worth the effort.

The intermediate cooling cycle could be studied by the methods which were demonstrated by John Erb of the Works Laboratory, General Electric Co., Schenectady, in an article entitled "Extending the Jominy Scale", which appeared in *The Iron Age* for June 13, 1946. Still slower

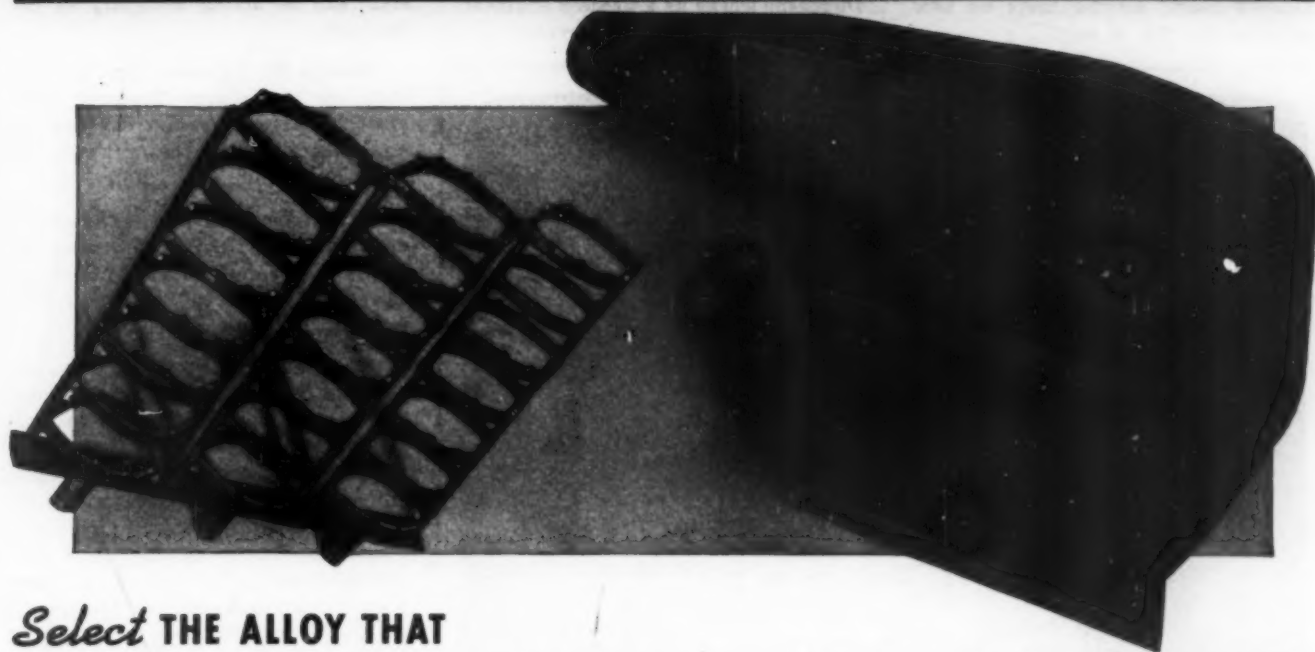
rates could be studied with the aid of a dilatometer synchronized with respect to the time-temperature changes in a suitable furnace. If a sufficient number of specimens were charged in the furnace, observation of the temperature-dilation curve would indicate precisely when specimens should be withdrawn from the furnace and quenched in order to "freeze" the structure at the proper time. The synchronization, of course, would be achieved by means of a program control that would permit the investigation of any arbitrary number of increasingly slow cooling rates, until practically any conceivable range of cooling transformations had been studied and charted, so as to guide the practicing metallurgist through some of his most perplexing problems.

Fig. 12—End of Isothermal Reaction in the Bainite Range for S.A.E. 4340 and 4360 (U.S. Steel's Atlas) With Author's Micros at 1000×



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Personals

John J. McClanahan ☉, upon completion of requirements for his B.S. at the University of Alabama, accepted the position of metallurgical engineer with the Alabama Pipe Co., Anniston, Ala.

Leroy C. Land ☉ has been assigned to the Army Field Forces Board No. 1 at Fort Bragg, N. C., where he will conduct users' service tests on new material for the Army.

Wallace F. Ardussi ☉, formerly with Industrial Rayon Corp. and Hupp Corp., has purchased the Variety Machine & Stamping Co., Cleveland.

Willard L. Groene ☉, formerly with Toulmin & Toulmin, is now practicing patent law and engineering in Phoenix, Ariz.

David M. Moses ☉, a June 1948 graduate from Virginia Polytechnic Institute, is now employed by the Carnegie-Illinois Steel Corp. at its Duquesne works as a process engineer in the metallurgical department.

A. B. Parsons has resigned after 18 years as secretary of the American Institute of Mining and Metallurgical Engineers in New York City and will move to Oakland, Calif. He is the eighth to hold the position of secretary for this 77-year-old organization, his surviving predecessors being Bradley Stoughton (past-president ☉) and for many years head of the metallurgy department at Lehigh University) and Frederick F. Sharpless. Pending election of a successor, Edward H. Robie of the A.I.M.E. staff will be acting secretary of the Institute.

Donald W. Frommer ☉, formerly with the Aluminum Research Laboratories as a corrosion engineer, is now a metallurgical engineer with the U. S. Bureau of Mines, Rolla, Mo.

A. L. Ascik ☉ has resigned his position with H. H. Robertson Co. and is now connected with Universal-Cyclops Steel Corp., Bridgeville, Pa., as superintendent of melting.

The United States Steel Corp. announces the following organizational changes: J. Norman Quinlan ☉ from division superintendent of the west mills of the Gary works of Carnegie-Illinois Steel Corp. to assistant to the general superintendent of the plant; Charles J. Hunter ☉ from chief metallurgist and inspector of the same plant to Mr. Quinlan's old position of division superintendent of the west mills; Oscar Pearson ☉ to chief metallurgist and inspector of the Gary works from assistant to division superintendent of steel production and central mills.

The Lake Erie Engineering Corp. of Buffalo, N. Y., announces that it is retaining the services of Gregory J. Comstock ☉ as consultant in the field of powder metallurgy. Mr. Comstock is also director of the powder metallurgy laboratory of Stevens Institute of Technology.

Vernon Pingel ☉, formerly senior research metallurgist with the Cornell Aeronautical Lab., is now head of metal powder products research and development at the Fansteel Metallurgical Corp., North Chicago, Ill.

Recently graduated from Michigan College of Mining & Technology, James C. Mallory ☉ is now employed as metallurgical engineer with the Ford Motor Co., Iron Mountain, Mich.

After receiving his M.S. from Stanford University, Richard C. Barry ☉ has accepted a position as assistant metallurgist at the Pearl Harbor Naval Shipyard.



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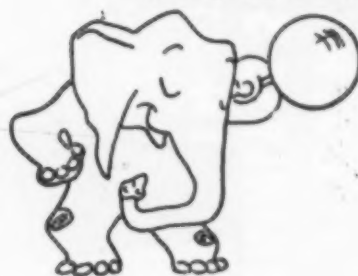
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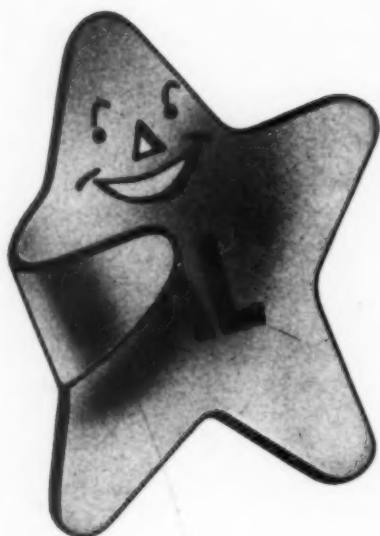
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Personals

James S. Stevenson, who graduated with a B.S. in metallurgy from Pennsylvania State College in June 1948, is now employed as a student engineer at the Campbell works of the Youngstown Sheet and Tube Co.

Resigning from Glenn L. Martin Co. where he was senior materials and process engineer, Harold J. Corwin has joined the Health Dept. of the State of Tennessee as a construction engineer.

Robert Hochscheid, who has recently graduated from the Colorado School of Mines, is now a junior engineer for the Dorr Co., of New York City.

Having received his Ph.D. from the State University of Iowa, Lewis W. Gleekman has accepted a position as assistant professor of chemical engineering at the University of Delaware.

E. R. Freeman is now employed by the Permanente Metals Corp., in the X-ray division of research, in Trentwood, Wash.

Albert A. Haskell, Jr., is now employed in the development engineering department, titanium division, National Lead Co., at South Amboy, N. J.

Following graduation from Ohio State University, Clinton C. McBride has accepted a position as research engineer at Battelle Memorial Institute, Columbus, Ohio.

Horace J. Weymer, after receiving his M.S. from Pennsylvania State College, has joined the American Brake Shoe Co.'s laboratory at Mahwah, N. J., in the capacity of assistant metallurgist.

Following graduation from Texas Technological College, S. A. Brooks has returned to Hughes Tool Co. as methods analyst in the industrial engineering department.

Graduating from Ohio State University in June 1948, Robert V. Steenstrup is now training in the scientific program of the General Electric Co. in the laboratory of the West Lynn works, Lynn, Mass.

Ralph Hodil, formerly with Youngstown Sheet & Tube Co., Good-year Aircraft Corp., and Surface Combustion Corp., has entered the consulting field as a metal coating specialist, with headquarters in Vandergrift, Pa.

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	Type 310, or stainless clad, or stainless to mild steel.	Chromend HCN Stainlend HCN
	Similar to Type 310 for welds requiring columbium or molybdenum.	Chromend 25/20Cb Chromend 25/20Mo
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	Type 316 Columbium Stabilized	Chromend KMoCb Stainlend KMoCb
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	Heat and scale resisting alloys —high temperatures.	Chromend 13/60 Chromend 14/75 Chromend 20/80 Chromend 15/85
STRAIGHT CHROME (Ferritic Alloys)	Type 410	Chromend 12
	Type 430	Chromend 16
	Type 442	Chromend 18
	Type 446	Chromend 28
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Personals

Donald E. Feather ☉, formerly foundry instructor in the department of mining and metallurgy at Wisconsin University, is now a trainee at Albion Malleable Co., Albion, Mich.

L. F. Dobyns ☉ is now supervisor of specifications at the Geneva Steel Co., Salt Lake City, Utah.

Following graduation from Pennsylvania State College, William H. Myers ☉ is presently employed in the research department of the Bethlehem Steel Corp., starting the "loop" course.

Eckard Hollister ☉, formerly with Oakite Products, Inc., is now sales representative in Eastern New York for Frederick Gumm Chemical Co.

Formerly a research fellow at Rensselaer Polytechnic Institute, R. D. Malin ☉ is now a trainee with American Manganese Steel Division of American Brake Shoe Co., St. Louis, Mo.

William C. Leprieh ☉, who was formerly with Ekco Products Co., has accepted a position with Pioneer Gen-E-Motor Corp., Chicago, in charge of the heat treating department.

Alwyn H. King ☉, who received his degree from McGill University in May 1948, is now employed as process engineer at Carborundum Co., Niagara Falls, N. Y.

Nicolas J. Cindric ☉ is now with the National Tube Co., Gary, Ind., enrolled in its engineering training program.

Jacques R. LaPointe ☉, formerly a field engineer with the Koppers Co., has resigned to accept a fellowship, offered by Albany Felt Co., at the Mellon Institute, Pittsburgh.

Edwin B. Johnson ☉, formerly a metallurgist with the Kaiser-Frazer Corp., is now employed as metallurgist with the Cleveland Cliffs Iron Co. at Ishpeming, Mich.

Joseph Ceder ☉ is now with Gitz Engineering as a shop foreman.

H. M. Pfahl ☉, formerly service metallurgist for Carnegie-Illinois Steel Corp., at Pittsburgh, is now sales engineer with U. S. Steel Export Co., New York City.

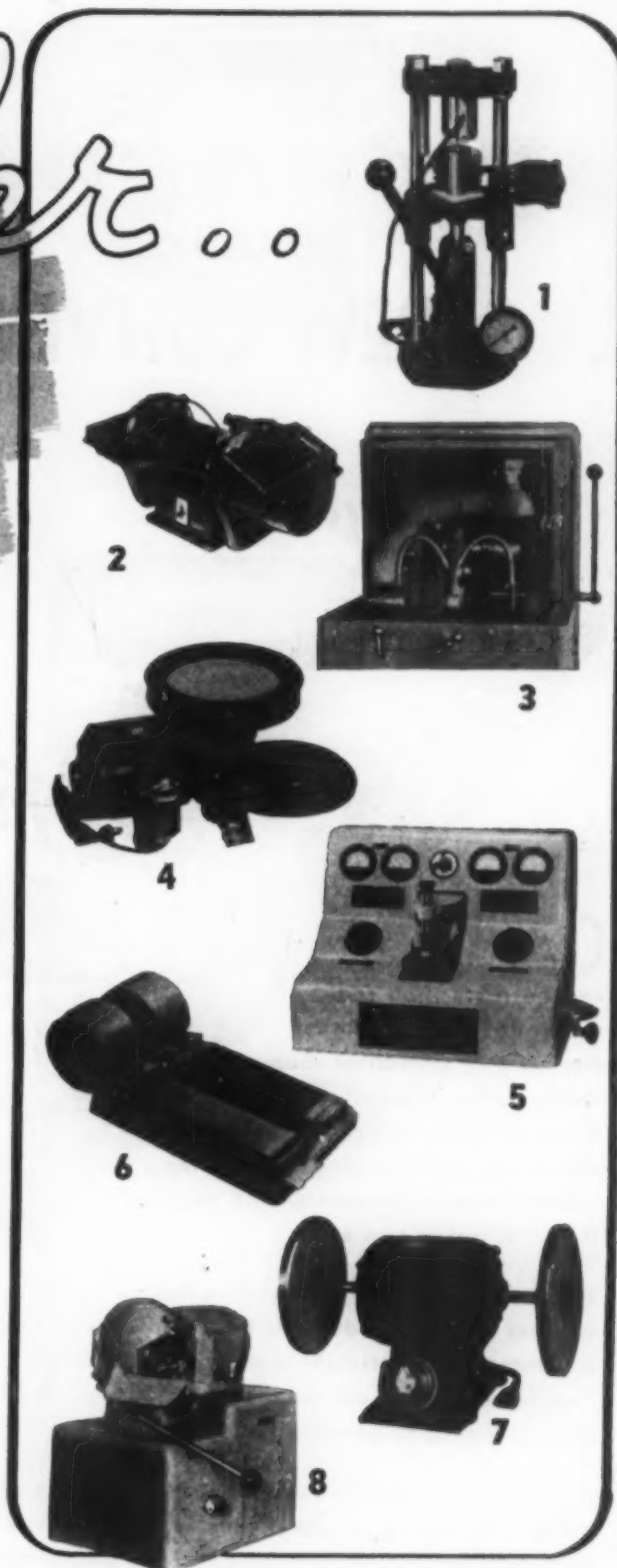
Walter B. Levy ☉, who graduated from Colorado School of Mines in May 1948, has accepted a position in the metallurgical laboratory of the National Supply Co., Torrance, Calif.

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Centrifugal Castings can be more than conventional piping. Almost any roughly cylindrical shape can be cast centrifugally provided a straight hole through the center is allowable. In the casting machine the molten metal is thrown outward, making it impractical to cast solid.

This Duraspun Screw Conveyor is typical of the unusual castings produced in our centrifugal casting department. On straight piping, our machines are capable of turning out pipes ranging in 2½" to 31" OD and, according to diameter, up to 15' long.

If you are interested in superior castings — more uniform, denser and pocket-free castings — order centrifugal castings. Write us about your requirements and we'll recommend the proper alloying elements to meet your corrosive, high temperature, abrasive conditions.

THE **DURALOY** COMPANY

Personals

R. C. Singleton ☉ has been promoted from chief metallurgist and technical adviser at Nelson Stud Welding Co., Lorain, Ohio, to head of customer engineering department.

Davis D. Wilson ☉ is now employed at the research and development laboratory of the New Mexico School of Mines, Albuquerque, N. M., as a research mechanical engineer.

J. Earl Romer ☉, for the past ten years district sales manager of Bliss & Laughlin, Inc.'s, Cleveland office, has been appointed district sales manager in Hartford, Conn.

Frank Kayser ☉, until recently employed in the metallurgy department of Unitcast Corp. of Toledo and Canadian Unitcast, Ltd., is now at Massachusetts Institute of Technology engaging in research and study sponsored by Vanadium-Alloys Steel Corp.

J. E. Micksch ☉ is with Lake City Malleable, Inc., Ashtabula, Ohio, as a research assistant.

Miles Bloom ☉, who graduated from Pennsylvania State College in June 1948, is now employed by the Youngstown Sheet and Tube Co., Youngstown, Ohio.

Rex Deghuée ☉, a past chairman of the St. Louis Chapter of the American Society for Metals, has been returned to the Bethlehem Steel Co.'s home office after 21 years in the St. Louis district office. He has been assigned to rail sales.

Eugene Cohn ☉ and his father, Sidney Cohn, have established the Secon Metals Corp. in New York City for the sale and processing of platinum metals and alloys.

Carl M. Carman ☉, who recently received his M.S. from Ohio State University and who worked for Ohio State University Research Foundation, has accepted a position with the research and development department of the Babcock & Wilcox Co. in Alliance, Ohio.

Theodore H. Gray ☉ has left the Westinghouse Research Laboratories to take the position of consulting metallurgical engineer with Metallurgical Engineers, Inc., Portland, Ore.

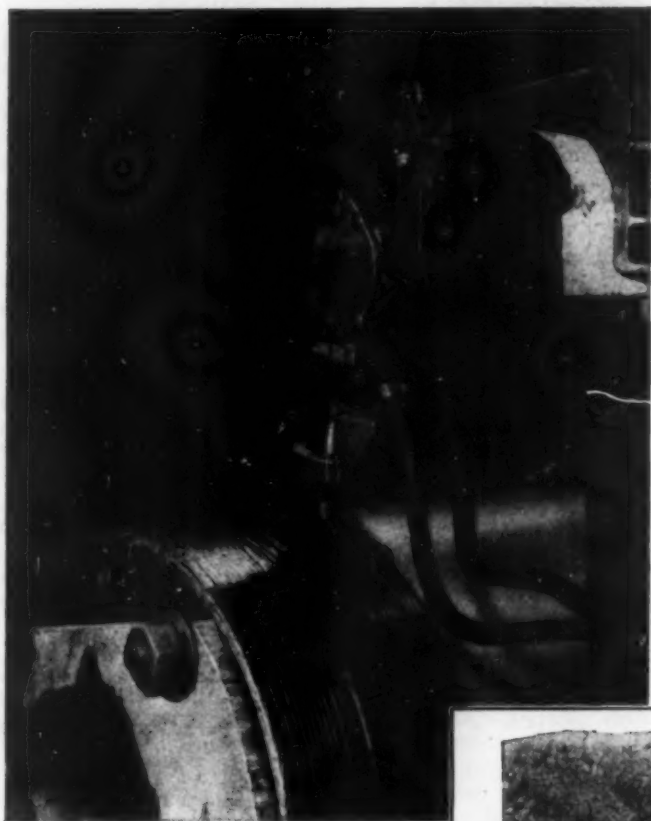
Having obtained a leave of absence from the Caterpillar Tractor Co., Raymond H. Hays ☉ has returned to the University of Kentucky to work for a M.S. in metallurgical engineering.

Give New Life to Worn Parts

. Rebuild Them by the **UNIONMELT Welding Process**

This automatic process saves time and money—welding is fast—parts are returned to service quickly—new surface is often superior to original—

Almost any part that can be suitably positioned can be rebuilt by the UNIONMELT welding process. It has been used successfully on parts such as steel mill rolls, guides, mandrels, journals; wheels for locomotives, cars, and cranes; pipe molds, and press plungers and cylinders.



Rebuilding a 24 in. diameter press plunger.

Deposits of almost any thickness can be made of—

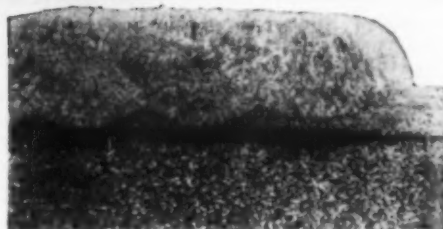
High carbon, flame-hardenable steels

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Corrosion- and heat-resistant steels

With the UNIONMELT electric welding process clean, dense, uniform weld metal can be deposited at the highest known welding speeds. There is no glare, flash, or spatter during welding. If a finished surface is needed, only minimum machining is required because of the smoothness of the weld deposit.

There are many LINDE methods of joining, rebuilding, forming, cutting and treating metals. LINDE Engineering Service can help you in production, construction, and maintenance. Just call the nearest LINDE office.




A flat surface is produced by offsetting each bead.

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High Precision Industrial Air Conditioning with extremely dry atmospheres (or with high relative humidities) at low temperatures.

● Specializing for thirty years in the more difficult problems of air conditioning for industrial processes, this Company has developed a group of units that make it easier and less expensive for you to get the particular air conditioning benefits you may need for your special process, or to overcome some obstacle of climate or condition that is interrupting your production or causing loss from rejected parts or materials.

The Niagara "No-Frost" Method, for example, has been used to create temperatures as low as -90°F . in cold test rooms, and to provide air with only 1 gr. of moisture per lb. for special processing.

The Niagara Type "A" Air Conditioner creates any condition of temperature and humidity for a test or process, and if wanted, creates different conditions in different rooms simultaneously.

Some of the industrial applications of these units: internal combustion engines, motors and air craft, super-chargers and carburetors, gas cooling and controlled atmosphere process, film, plastics, fiber, rubber and adhesives control, biological processing such as penicillin, and yeast.

Write for a Niagara Blower Bulletin on a subject which interests you, or for the address of the nearest Niagara Field Engineer.

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HUMIDIFYING • AIR ENGINEERING EQUIPMENT		

Personals

LeRoy H. Markway ☉ is now employed with the Wood River, Ill., refinery of the Shell Oil Co., as a junior engineer.

Lewis D. Conta ☉ has left Air Reduction Research Laboratories at Murray Hill, N. J., to become professor of mechanical engineering at the University of Rochester.

Henry E. Kiley ☉, formerly on the engineering staff of Standard Machinery Co., has recently become associated with Universal Winding Co. of Providence, R. I., as research engineer.

Frank J. Mehringer ☉ has left the faculty of Massachusetts Institute of Technology to join Victory Plastics Co., Hudson, Mass., as research and development engineer.

Ronald L. Vingoe ☉, formerly a research engineer at Battelle Memorial Institute, is now associated with the Pittsburgh Forgings Co. of Coraopolis, Pa., as a metallurgical engineer.

Charles J. Swan ☉, having graduated from Cornell University, is now participating in the training program of the Standard Oil Co. of Ohio, in Cleveland.

Robert L. Ray ☉ is now with the new Joseph T. Ryerson plant in Emeryville, Calif., as metallurgist of the special steels department.

Having recently graduated from the University of Utah, Ivan L. Nichols ☉ has been employed as assistant metallurgist at the Pacific Bridge Co., Park City, Utah.

R. Wayne Parcel ☉ has recently been appointed assistant professor of metallurgy at Carnegie Institute of Technology.

Joseph A. Bowman ☉ has been transferred to the Rochester, N. Y., office of Brown Instrument Co. from the Syracuse office.

William F. Hofmeister ☉ is now research metallurgist at Chain Belt Co., Milwaukee, Wis.

Upon graduating from the Michigan College of Mining and Technology, Rodney P. Roehm ☉ has accepted employment with the Wyman-Gordon Co. in Worcester, Mass.

Ulf Landergren ☉, on leave from the Electrical Welding Co. of Gothenburg, Sweden, is now studying at Carnegie Institute of Technology, Pittsburgh, toward his doctor's degree.



(Above) New 5,000,000 lb. machine at left. At right is a 600,000 lb. Baldwin machine.

The largest universal testing machine in the world, built to Navy specifications by Baldwin, is now at work in the Aeronautical Structures Laboratory at the Philadelphia Naval Base. It is capable of applying loads of 5,000,000 pounds in tension and compression, on specimens up to ten feet wide and thirty feet long. Flexure loads may be applied to structures up to 50 feet long.

Initial assignment of the new machine is in the field of aircraft development. Full-sized simulated aircraft structural members and assemblies of aluminum and magnesium will be tested. The aim is to achieve the optimum in shape, weight and strength and also to develop factors to take care of "size effect," thus permitting more trustworthy projection of test-specimen results, and

making model testing a more accurate basis for design.

Engineers today are increasingly reluctant to assume direct correlations between tests of small specimens and performance of large sections or structures, and are demanding verifying evidence—which can only be provided by large capacity machines. To meet this requirement, Baldwin has built and installed eight other universal testing machines, with capacities of 1,000,000 to 4,000,000 pounds. The effectiveness of these new tools has been magnified many times through use of the Baldwin SR-4* strain gage, which accurately reports the stress pattern in the part under study.

If full-scale testing has become an essential in your business, Baldwin will be glad to discuss your problem with you.

The Baldwin Locomotive Works, Philadelphia 42, Pa., U.S.A.
Offices: Boston, New York, Philadelphia, Houston, St. Louis,
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ington. In Canada: Peacock Brothers, Ltd., Montreal, Quebec.

*Trade Mark Registered U. S. Patent Office

TESTING HEADQUARTERS **BALDWIN**



(Above) "Nerve center" of the new testing machine. Large dials show load.

(Right) Eccentric load test produced only a fraction of allowable deflection.

(Below) An 8-in. test specimen is snapped under a 4,200,000 lb. load.



"Miss Great Lakes" Gold Cup winner for '48**Depended on Wheelock, Lovejoy Alloy Steel**

Photo Courtesy The Detroit News

UNUSUALLY ROUGH RACING CONDITIONS DISABLED 12 OUT OF 14 BOATS

Properly selected alloy steel, from which vital equipment was fabricated, played a leading role in winning the grueling 1948 Gold Cup race that saw only 2 of 14 starters cross the finish line. The winner, "MISS GREAT LAKES", driven by Danny Foster and owned by Al Fallon, had underwater struts, rudder, jack shaft and steering pitman arm made from Wheelock, Lovejoy HY-TEN B #3X steel. This particular alloy was selected and properly heat treated to meet the most rugged conditions possible. The fact that not one of these parts failed or even bent, in spite of the terrific beating they took, is ample proof of HY-TEN's superior physicals. And it's proof too that Wheelock, Lovejoy knows steel. Perhaps you have a tough job that demands just the right steel—write Wheelock, Lovejoy today.

WL steels are metallurgically constant. This guarantees uniformity of chemistry, grain size, hardenability—thus eliminating costly changes in heat treating specifications.

Write today for your **FREE COPY** of the Wheelock, Lovejoy Data Book. It contains complete technical information on grades, applications, physical properties, tests, heat treating, etc.



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BILLETS AND FORGINGS FOR PRODUCTION. TOOL ROOM AND MAINTENANCE REQUIREMENTS.

Personals

Following graduation from the Pennsylvania State College, Owen A. Ridenour ☉ has accepted a position in the quality control department of the Steubenville, Ohio, works of the Wheeling Steel Corp.

Gilbert P. Muir ☉, formerly editor of "Watkins Cyclopedia of the Steel Industry" and of *Steel Processing*, has been appointed editor of the *Tool Engineer*, official publication of the American Society of Tool Engineers.

Formerly metallurgist for the Firestone Tire and Rubber Co., Carl M. Erb ☉ is now a vendor contact metallurgist for the Kaiser-Frazer Corp., Willow Run, Mich.

Robert B. Smock ☉ has transferred from the special projects section, aviation ordnance department, Naval Proving Ground, Dahlgren, Va., to the instrumentation section, technical development department, Naval Aviation Ordnance Test Station, Chincoteague, Va.

George A. Risko ☉, who graduated in September 1948 from the school of mines, University of Pittsburgh, is presently employed by the National Tube Co. as a junior metallurgist.

David A. Elmer ☉, who was formerly with the California Institute of Technology, is now with C. F. Braun & Co.'s research department in Alhambra, Calif.

Jos. B. Clough ☉ has recently become associated with the A. W. Hecker Co., Cleveland, as assistant to the president.

John J. Uppgren ☉ is now instructor in foundry practice at the University of Minnesota.

Wayne L. Cockrell ☉, formerly senior metallurgist at Oak Ridge National Laboratories, is now a project engineer with the Kellogg Corp., New York City.

After receiving his M.S. degree from the Pennsylvania State College in June 1948, John F. Beck ☉ is now associated with the Babcock & Wilcox Tube Co., Beaver Falls, Pa., as a research metallurgist.

Wm. Hawkesworth ☉ has recently changed positions from the technical service department, Bridgeport Brass Co., to metallurgist, Gibbs and Cox, Inc., naval architects, New York City.

T. W. Sproull ☉ is now employed by the Babcock & Wilcox Co. at Barberton, Ohio.



Products for the Iron and Steel Industries

PRODUCT	TYPICAL COMPOSITION	APPLICATIONS
ALSIFER	Aluminum.....20% Silicon.....40% Iron.....40%	Used principally as a steel deoxidizer and for grain size control.
FERRO CHROMIUM High Carbon Grade	Chromium....66-70% Carbon.....4-6%	For wrought constructional steels and steel and iron castings.
Iron Foundry Grade	Chromium....62-66% Carbon.....4-6% Silicon.....6-9%	For alloyed cast irons. Readily soluble as a ladle addition at the lower temperatures of cast iron.
Low Carbon Grades	Chromium....67-72% Carbon... .06%, .10%, .15%, .20%, .50%, 1.00% and 2.00% max.	For low carbon chromium steels, especially those with high chromium content, such as stainless and heat-resistant types.
FERRO SILICON 25-30% Grade	Silicon.....25-30%	Deoxidizer for open hearth steels; also for high silicon, corrosion-resistant iron castings.
50% Grade	Silicon.....47-52%	Used as a deoxidizer and for the addition of silicon to high silicon steels, for springs, electrical sheets, etc. Pulverized form used as ladle addition to cast irons for silicon content and graphitization control.
75% Grade	Silicon.....74-79%	For high content silicon steels, such as spring steels, sheets and forgings of high magnetic qualities for electrical apparatus.
High Silicon Grades 80-85% 85-90% 90-95%	Silicon....80-84.9% Silicon....85-89.9% Silicon.....90-95%	For high content silicon steels where small ladle additions are used for required silicon content. Also for manufacture of hydrogen by reaction with caustic soda and production of magnesium by the Pidgeon process.
FERRO TITANIUM High Carbon Grade	Titanium....15-18% Carbon.....6-8%	Final ladle addition to control "rimming" action and to clean effervescing steel. Final deoxidizer and scavenger for steel castings and fully killed ingot steels.
Medium Carbon Grade	Titanium....17-21% Carbon.....3-4.50%	Often preferred to the High Carbon Grade as a final ladle addition to very low carbon rimming or effervescing steels.
Low Carbon Grades 20-25% Ti Grade	Titanium....20-25% Carbon...max. 0.10% Silicon.....max. 4% Aluminum max. 3.50%	Carbide stabilizer in high chromium corrosion-resistant steels of extremely low aluminum content. Deoxidizer for some casting and forging steels.
40% Ti Grade	Titanium....38-43% Carbon...max. 0.10% Silicon.....max. 4% Aluminum...max. 8%	Carbide stabilizer in high chromium corrosion-resistant steels, where smaller ladle additions are desired and extremely low aluminum content is not essential.
VANADIUM METAL 90% Grade	Vanadium....91.15% Aluminum....2.25% Silicon.....0.50% Carbon.....0.17%	For special iron-free (non-ferrous) or low iron alloys or low impurity ferrous alloys.
95% Grade	Vanadium....95.18% Aluminum....2.00% Silicon.....0.27% Carbon.....0.40%	Principally for research on the properties of pure alloys, where very low iron content is essential.

PRODUCT	TYPICAL COMPOSITION	APPLICATIONS
FERRO VANADIUM Grade "A" (Open Hearth)	Vanadium....35-45% 50-55% Silicon....max. 12% Carbon...max. 3.50%	For low percentage vanadium content of rolled, forged or cast constructional steels. Also used in vanadium cast irons.
Grade "B" (Crucible)	Vanadium....35-45% 50-55% Silicon...max. 3.50% Carbon...max. 0.50%	For tool steels and special steels containing high percentages of vanadium, in which required limits for carbon and silicon are narrow.
Grade "C" (Primes)	Vanadium....35-45% 50-55% 60-65% 70-80% Silicon...max. 1.25% Carbon...max. 0.20%	For tool steels and special steels requiring high percentages of vanadium and exceptionally low carbon and silicon content.
VANADIUM PENTOXIDE Technical Grade Fused Form	V_2O_590%	A source of vanadium in basic electric furnace steels. A base for numerous chemical compounds.
Technical Grade Air Dried Form	V_2O_583-85%	A base for preparation of numerous chemical compounds (catalysts, etc.).
GRAINAL ALLOYS Vanadium Grainal No. 1	Vanadium....25.00% Aluminum....10.00% Titanium....15.00% Boron.....0.20%	Practical and economical intensifiers for controlling and increasing the capacity of steels to harden, and for improving other important engineering and physical properties.
Vanadium Grainal No. 6	Vanadium....13.00% Aluminum....12.00% Titanium....20.00% Boron.....0.20%	See above.
Grainal No. 79	Aluminum....13.00% Titanium....20.00% Zirconium....4.00% Manganese...8.00% Boron.....0.50% Silicon.....5.00%	See above.
GRAPHIDOX No. 4	Silicon.....42-46% Titanium.....9-11% Calcium.....5-7%	For graphitization of iron; ladle treatment insures normal graphite, free from dendritic structure; reduces chill; efficient inoculant in production of high strength irons.
V-FOUNDRY ALLOYS V-5 Grade	Chromium....38-42% Silicon.....17-19% Manganese...8-11%	Used in cast irons as a ladle addition. Reduces chill, promotes uniformity of structure, increases strength and hardness.
V-7 Grade	Chromium....28-32% Silicon.....15-21% Manganese...14-16%	See above.
BRIQUETTES Ferro Chromium	Hexagonal. Weigh approx. 3 1/2 lb. and contain 2 lb. of chromium.	A practical and convenient form for adding ferro-alloys to the cupola.
Ferro Silicon	Two sizes, both cylindrical, one containing 1 lb. of silicon; the other, 2 lb. of silicon.	See above.
MISCELLANEOUS	Special ferro-alloys, metals, chemicals and carbides.	To meet individual requirements.

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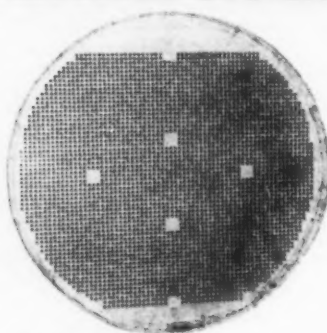
PITTSBURGH

SMALL METAL TUBING
(Max. O.D. $\frac{3}{8}$ ")

MUST WITHSTAND PRESSURE AND HEAT



**CONDENSERS
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There's nothing like knowing—in advance—that *Superior* tubing possesses the inherent qualities of safety, endurance and reliability so necessary when the tubing is to be used to contain liquids or gases *under pressure*. Superior tubing in carbon and stainless steels is produced to respond excellently to bending, flaring, flanging, pulsating, etc.—made with relatively high endurance characteristics that eliminate early tube failure.

With Superior tubing you get *all* of these qualities—

- ... Straight, clean, bright tubing—for smooth flow of media.
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- ... Maximum, uniform softness—for ease of bending and flaring.
- ... Standardization under current ASTM specifications.

We welcome the opportunity to furnish you with full information regarding your particular tubing requirements. Bulletin #31, listing analyses available and commercial tolerances, will be sent upon request.

We will be glad to see you at our Booth #1829 at the National Metal Exposition.

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Personals

Following graduation from the University of Illinois, **Joseph B. Selle** has joined the Ohio Steel Foundry Co., Springfield, Ohio, as a sales engineer.

Stuart M. Hansen, who graduated from Michigan College of Mining and Technology in June 1948, is now located at the Lackawanna, Pa., plant of the Bethlehem Steel Co., training in the "loop" course.

Edward A. Erickson has resigned from Battelle Memorial Institute to accept a position as senior combustion engineer in the Sears, Roebuck Mechanical Testing and Development Laboratory, Chicago.

Edward W. Roath has been promoted from assistant general superintendent of Kennecott Wire & Cable Co., Phillipsdale, R. I., to superintendent of the company.

I. Jordan Kunik recently joined the patent department of International Telephone & Telegraph Corp., New York City, as a patent attorney.

Frank J. Ross is a research chemist in the U. S. Naval Research Laboratory, High Polymer Section, Washington, D. C.

R. J. Zale has recently been transferred to the Rochester division of the Lindberg Steel Treating Co., where he is plant metallurgist.

George Reed, who was formerly sales engineer for the Steel and Tube Division, Timken Roller Bearing Co., at Houston, Texas, is now chief inspector at the Berwick, Pa., plant of the American Car and Foundry Co. and assistant manager of inspection for the eastern plants of the company.

George J. Salvaggio, who recently received his master's degree from Carnegie Institute of Technology, is now employed as research engineer in the research and development department of Jones & Laughlin Steel Corp., Pittsburgh.

William L. Frankhouser has been recently employed by Babcock & Wilcox Tube Co. in Beaver Falls, Pa., as a student in its engineer training program.

Donald E. Nulk has accepted a research assistantship at Carnegie Institute of Technology, where he will continue his graduate studies and work at the Metals Research Laboratory on an Atomic Energy Commission project.

High Temperature Lubrication...400°F. and up!

Clean parting of castings from permanent molds is an important demand of efficient foundry operation. The Arrow Aluminum Castings Company, Cleveland, Ohio, gets better mold reproduction, longer mold life, easier parting and smoother surfaces on its castings by coating permanent molds with "Sono-Graph", a mold wash produced with "dag" colloidal graphite by the Mealey Grease and Oil Company, Cleveland, Ohio.

In countless other industries too...wherever high temperature lubrication is a problem...one of the many "dag" colloidal graphite dispersions is the likely answer. This extremely versatile substance can also be used in opaquing, in general lubrication, in impregnating and in electronics work.

Acheson Colloids engineers may be able to help you with your problems. Clip the coupon and mail now.

Acheson Colloids Corporation, Port Huron, Michigan; Boston; New York; Philadelphia; Pittsburgh; Cleveland; Detroit; Chicago; St. Louis; Los Angeles; San Francisco; Toronto.



Pouring into permanent
molds at the Arrow Aluminum
Castings Company, Cleveland, Ohio

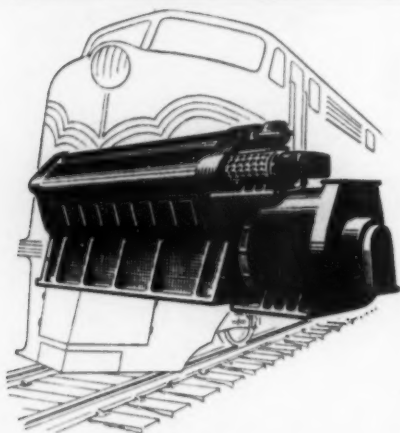
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Diesel engines in railroad locomotives are piling up an enviable trouble-free performance record. Many castings in these engines contain specified Molybdenum contents.

When wear or heat must be resisted, and high strengths are required, the use of Molybdenum in cast iron is becoming more and more routine.

The listing of a cylinder liner for Diesel use in railroad locomotives containing 3.20% Carbon, 1.90% Silicon, 0.85% Manganese, with 0.20% Chromium, 0.40% Molybdenum and 0.45% Nickel added can be found on page 10 of our new booklet—"Applications of Molybdenum Cast Irons."

Dozens of other alloy cast iron compositions used in Diesel engines, as well as hundreds of other Molybdenum-containing compositions are similarly listed in this new booklet. Write for it!

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"Applications of
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F9

Personals

Valore L. Marcinak is now vice-president-treasurer of Tech Agency, an advertising agency specializing in metallurgical products.

Anthony S. Rugare is presently engineer in charge of germanium diode production at the General Electric Co., Syracuse, N. Y.

Cecil J. Bier is now chief engineer and partner of Powmetco, Inc., which has moved its entire plant from Pittsburgh, Pa., to East Port Chester, Conn.

Harry K. Ihrig has been elected vice-president and director of laboratories of Globe Steel Tubes Co. Dr. Ihrig has been with the company since 1934.

R. Sidney French, research metallurgist for Bridgeport Brass Co. for the past ten years, has affiliated himself with Metallurgical Products Co., Bridgeport, Conn., as vice-president.

William J. Bronkala, Jr., a recent graduate of the University of Minnesota's school of mining and metallurgy, has joined Stearns Magnetic, Milwaukee, as an assistant physicist and research engineer.

John G. Dean has been appointed head of the industrial chemicals section, Development and Research Div., The International Nickel Co., Inc. He has been associated with the company since 1942 as a senior fellow in International Nickel's fellowship at Mellon Institute of Industrial Research and as an advisory fellow at the headquarters in New York.

Anthony R. Ciuffreda, formerly with Metal & Thermit Corp., has accepted a position on the engineering staff of the Standard Oil Development Co., Linden, N. J.

Wheeling Bronze Casting Co. announces the appointment of Harry P. Croft as vice-president in charge of development. Dr. Croft was formerly associated with the Chase Brass & Copper Co., Inc., and during the war he served as copper consultant for the Bureau of Conservation, Office of Production Management, and as chief of the Industrial Division, Cleveland Ordnance District. He was elected a trustee of the company at the annual meeting on Oct. 27, 1948.

Alfred Strasser, having graduated from Purdue University in June 1948, is now a junior metallurgist at the M. W. Kellogg Co.'s special projects division, Jersey City, N. J.

HIGH SENSITIVITY TEMPERATURE MEASUREMENT



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many applications of this

PORTABLE PRECISION POTENTIOMETER

Distinctive features include:

1. High sensitivity, sturdy, built-in Pointer-lite galvanometer—permits balancing to within 2 microvolts in low-resistance circuits—better than 0.05° C. on iron-constantan couples.
2. Completely self-contained assembly—no external accessories except the thermocouple circuit.
3. Two full-scale ranges—0 to 16.1 millivolts and 0 to 161 millivolts—readable to within 2 and 20 microvolts respectively.
4. Convenient arrangement of galvanometer scale, potentiometer dials, keys and battery rheostats for greatest ease in reading and adjustment.
5. Sturdy, compact construction for many years of service under hard use.

Portable Precision Potentiometers are available in a selection of ranges up to 1.6 volts. Described with other Rubicon potentiometers in Bulletin 270 and 270-A.

OTHER RUBICON PRODUCTS

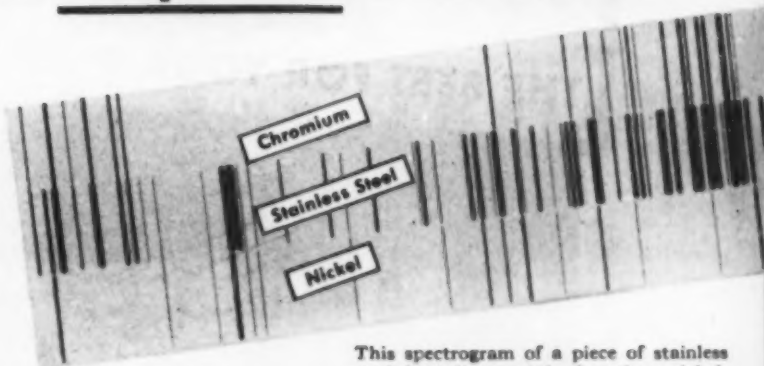
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testers for production testing • Evelyn
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electrical quantities.

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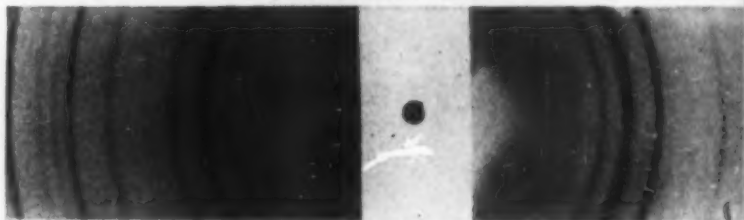
3 ways photography shows composition of materials

Spectrography . . . To record composition quantitatively and qualitatively, Kodak offers 108 specialized materials. For general spectrographic use, however, Kodak recommends two emulsions: *Kodak Spectrum Analysis No. 1 Plates and Films* for high contrast, low background density, and low granularity; and for a more uniform contrast-wavelength relation, *Kodak Spectrum Analysis No. 2 Plates and Films*.



This spectrogram of a piece of stainless steel shows it to contain chromium, nickel, and iron, but none of the other alloying elements commonly found in commercial stainless steel.

X-ray diffraction . . . To record crystalline composition, stresses, and thermal history by x-ray diffraction, the fastest film to use is *Kodak Industrial X-ray Film, Type K*. When smooth microdensitometer traces rather than highest film speed are required, use fine-grain *Kodak Industrial X-ray Film, Type A*.



This x-ray diffraction pattern shows, by the density variations along the Debye-Scherrer rings, notably by the patchy center, that the stainless steel used above is not isotropic in its properties.

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This electron diffraction study (below) of the surface of the same stainless steel shows the presence of a thin layer, formed when the steel was treated with concentrated nitric acid for greater resistance to corrosion.

Kodak will send you free: a copy of the booklet "Materials for Spectrum Analysis" and a chart that tabulates Kodak films for x-ray diffraction. Your specific questions on Kodak Lantern Slide Plates for electron diffraction, and on other products mentioned above, will gladly be answered by correspondence.

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Recent Metallurgical Progress in France

By M. G. Corson

FRANCE was occupied by the Germans from June 1940 to August 1944—almost to the end of the war. Knowing how disdainfully Hitler looked upon his adversaries, one might have expected to hear that the Germans reduced the French to a position only a little better than the one left to the Poles. One would have expected the French blast furnaces to work strenuously under the German whip and French openhearth to run large amounts of ingots or billets to be processed further by the only noble race. One would have expected the French metallurgists to be engaged in the nearly menial tasks of chemical testing and simplest physical checking of the intermediate products.

The real situation appears to be quite different. Scientific work done by the French metallurgists in 1942 to 1945 was at least as good and thorough as it was before the war—in my own opinion it was much deeper and carried more precision than the work done before, let's say, 1934, when I made my latest visit to France.

I would not dare to pass moral judgment on our French colleagues, even though it is quite obvious that their painstaking investigations greatly benefited the common enemy—but I must state the very unpleasant truth, that the Germans did not treat the French metallurgists as enemies. Quite to the contrary, they encouraged the Frenchmen to do substantial scientific work and granted them even enough paper to publish their results. It was kind of them!

The following represent pale abstracts of some of the substantial papers published in the *Revue de Métallurgie* for 1944.

Toughness

THE center of gravity of a paper by J. Pomey ("A Contribution to the Study of Fragility", p. 17, 49 and 83) lies in the problem of impact resistance. His preliminary tests on three types of structural alloy steel (one Ni-Cr and two Cr-Mo steels with carbon around 0.40%) have shown that no treat-

(Continued on p. 876)



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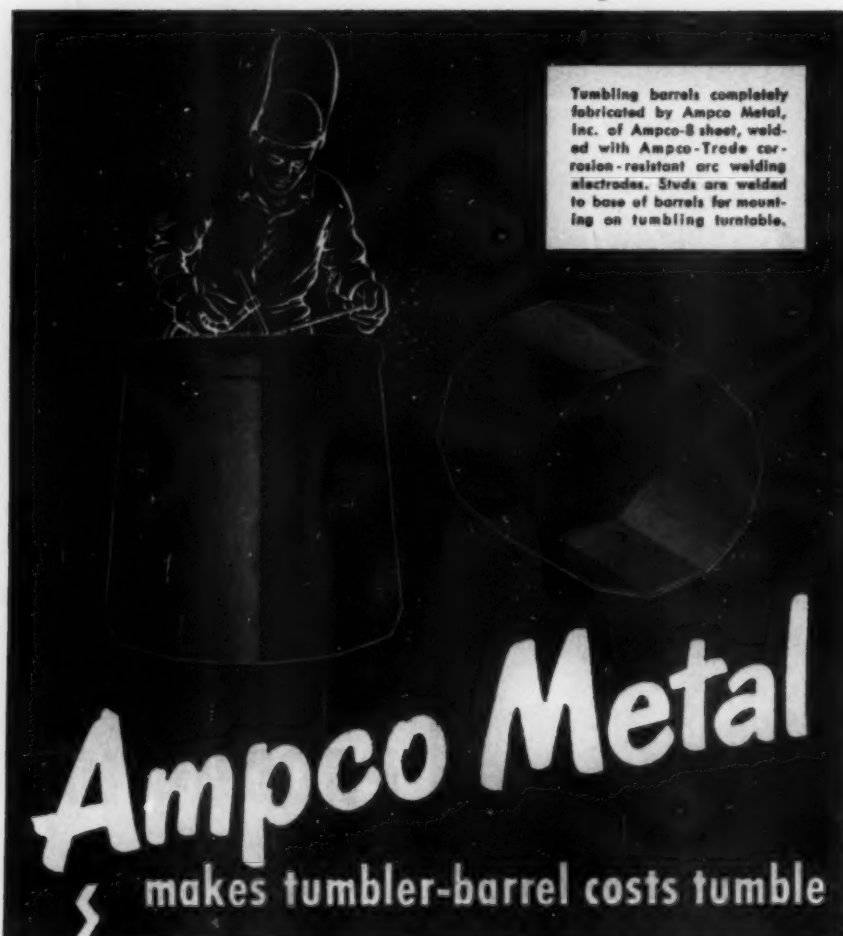
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Toughness

(Continued from p. 874)

ment can result in a product combining both a maximum endurance and a maximum resilience (impact resistance). This is because the maximum endurance, obtainable by tempering near 350° C. (660° F.), is accompanied by a minimum of resilience—while the maximum of the latter, obtainable at around 650° C. (1200° F.), is connected with a rapid decline in the absolute endurance value. On the other hand this maximum of resilience is followed by a high ratio of endurance to ultimate tensile strength.

Next Pomey investigated the characteristics of nine structural steels coming either from the open-hearth or the arc furnace. Test bars were taken both in the longitudinal and the transverse directions. In every steel the well-known fact that the ultimate tensile strength changes but little from one direction to the other found perfect corroboration. The endurance and resilience values differed, however, to a considerable extent—always in the same direction but not in the same amount. The drop in impact strength (using Mesnager's test bar) was always greater than the drop in endurance. On the other hand, certain of the steels—particularly samples taken from round billets—had nearly the same resilience and fairly close endurance values in both directions.

Pomey used Cazaud's formula

$$f_t = 7 + R_s \left[0.23 + 0.035 \sqrt{K} + 0.1 \left(\frac{K_T}{K_L} \right)^2 \right]$$

where f is the endurance limit in kg. per sq.mm.

R is the ultimate tensile strength in kg. per sq.mm.

K is the impact energy in kg-m. absorbed per sq.cm.

Subscript T represents transverse direction, and subscript L represents longitudinal direction.

In its reduced form this becomes

$$f = 7 + R (0.255 + 0.035 \sqrt{K})$$

Pomey computed the probable endurance value for a given steel in a given direction from the simplified formula, knowing R , the ultimate tensile strength, and K , the impact strength in that direction, and found the computed results to agree with the experimental results within 10%.

(Continued on p. 878)

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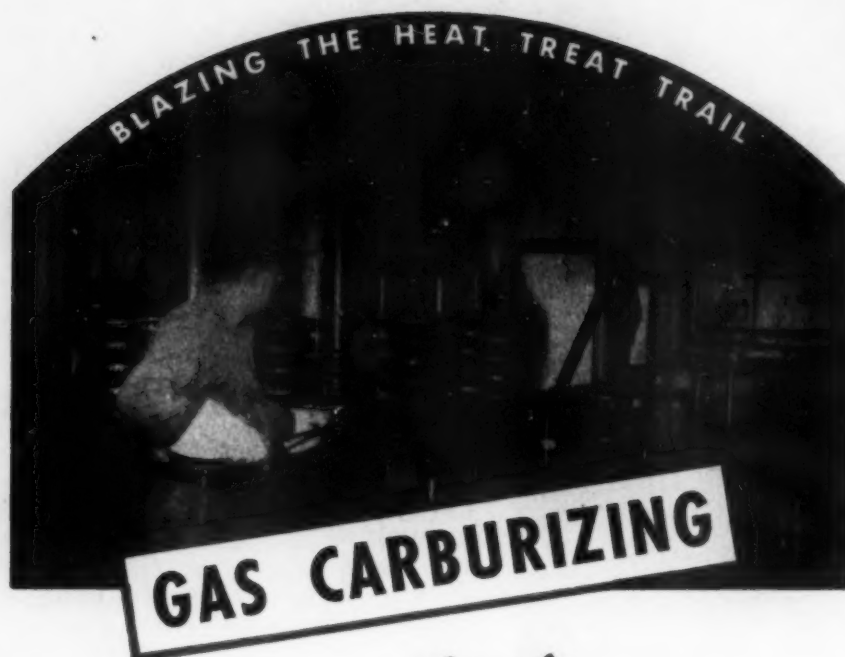
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Toughness

(Continued from p. 876)

The second part of Pomey's paper compared impact tests using various test bars. Hundreds of tests were made but the scientific results were not encouraging. No correspondence between the values obtained using the different test bars was found, nor was any one decidedly superior in the task of detecting quality differences. Pomey remarked that far better methods for a scientific approach to the study of impact strength can be devised.

In the third part of Pomey's paper the reader can find plenty of theoretical considerations regarding the pure stress of de-cohesion, intrinsic de-cohesion curves, and the Mohr circle of stresses. To understand that portion the reader must, however, become acquainted with the ideas of Caquot presented in the *Mémoires de la Société des Ingénieurs Civils de France* in October 1933, and with his book "Equilibre des Massifs à Frottement Interne" (Equilibrium of Massive Bodies Under Internal Friction), published in 1934.

Antifriction Alloys

THE problem of antifriction alloys is so complicated and so dependent on purely commercial considerations that substantial technical investigations are rarely undertaken or at least rarely published. An article in *Revue de Métallurgie* (1944), p. 78, by J. Lacombe and M. Dannenmuller—apparently the first of a series planned—forms therefore a considerable contribution to metallurgical knowledge.

The authors developed some special equipment for the study of the creep produced at constant temperature under different loads in a number of commercial alloys, frequently changing the composition slightly in order to examine the influence of the individual elements. All their work was done at 40° C., thermostatically regulated, and the progress of the creep was recorded photographically on a 50 × scale.

Their studies covered a total of 28 alloys, the basic metals being Sn, Pb, Cd, Zn, Cu, Mg and Al. Following Professor Portevin's suggestion that numbers are far more comparable than curves, especially of the semi-logarithmic kind, they

(Continued on p. 880)

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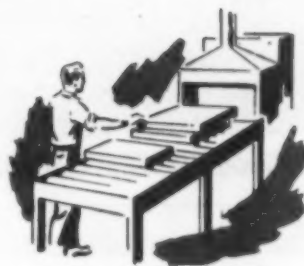
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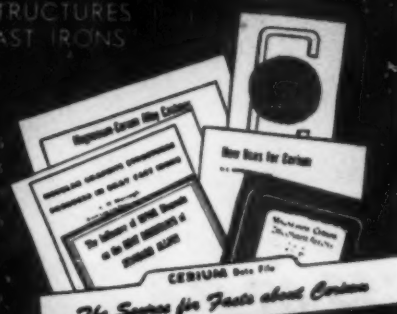

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Antifriction Alloys

(Continued from p. 878)

reduced their experimental results to sets of three factors, one representing the ratio between the immediate deformation on the application of the load and the total deformation in 200 hr., the other two being powers of time m and n , the first standing for the deceleration and the second for the acceleration of the creep with time. Their tables carry also the figure for the load per sq.mm. which would not produce more than 1% of the total compression in 200 hr.

Metallurgically interesting also is their statement that neither grain size nor the pouring temperature had any appreciable effect upon the creep of the alloys studied.

Coalescence of Carbide in Steel

"AN Introduction to the Study of Coalescence in Eutectoid and Hypereutectoid Steel" is the title of a series of articles in *Revue de Métallurgie* during 1944, written by Col. N. T. Belaiew. The author began his metallurgical investigations in the beginning of this century as a young artillery officer of the Russian army and a most promising young scientist on the staff of the Imperial Academy of Ordnance. As an army man he was quite naturally interested in the characteristics of such steels as were used in the manufacture of weapons, especially the fundamental weapon of many centuries—the sword. Hence his first studies were dedicated to those famous oriental "blades of Damascus". They were manufactured by mixing small chunks of iron of various carbon content—as was known from their relative hardness—including even small pieces of pig iron, heating the mass in a crucible under a cover of charcoal, welding it together at a moderate heat under the hammer, reheating, and forging again and again until a fine blade resulted (or the whole mass was thrown away if the operation were a failure).

Belaiew succeeded in proving 42 years ago that the specific properties of the steel of Damascus were due to the simultaneous presence of bands of ferrite and pearlite which gave it the miraculous toughness, while the presence of bands of iron

(Continued on p. 882)

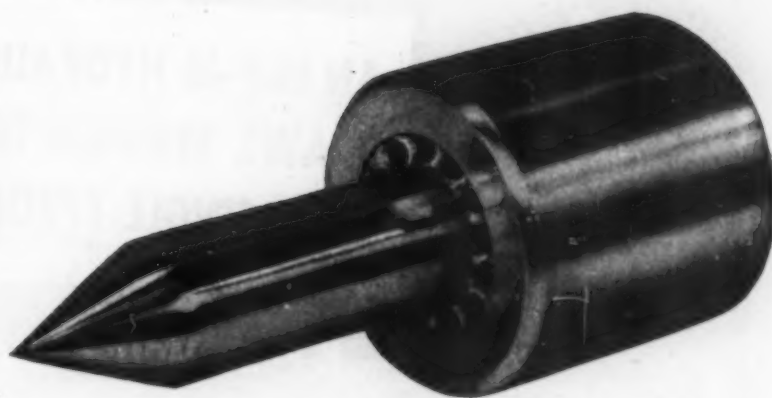
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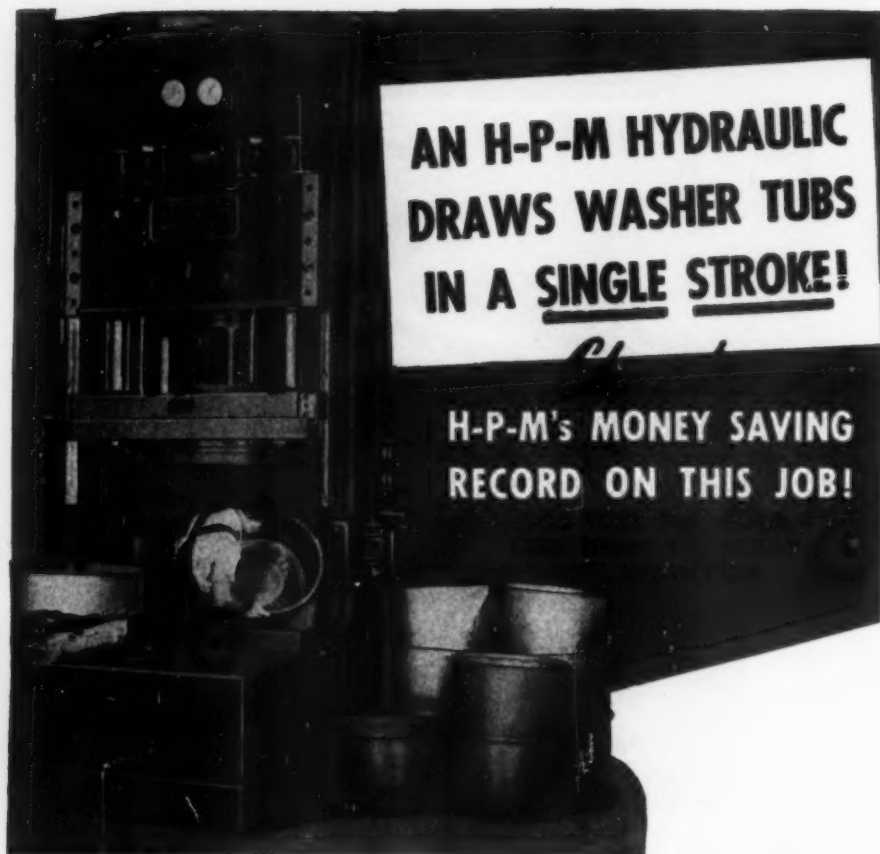
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Coalescence of Carbide

(Continued from p. 880)

carbides of considerable size gave it its cutting characteristics and acted as a regulating influence upon the changes in the pearlite during quenching and tempering.

Since these days he has been intensely interested in hypereutectoid steels and on numerous occasions brought forth the idea that for a large number of applications a plain carbon steel, properly treated, could well substitute for the more expensive alloy steels—a principle, it must be agreed, often put into practice by American metallurgists who are able to think for themselves.

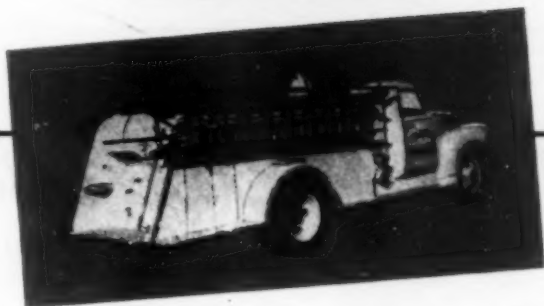
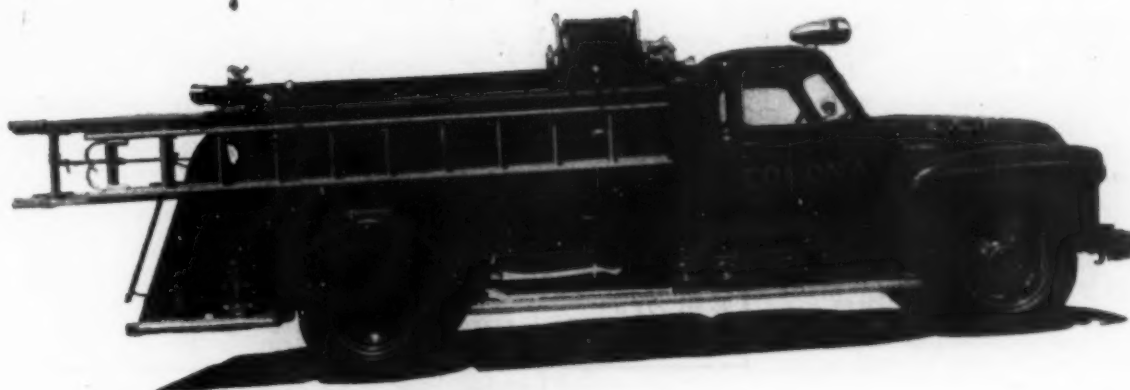
In the articles here considered, Colonel Belaiew gives a reasoned history of all research work done in the realm of pearlitic and hypereutectoid steels, as far as the sources were available to a scientist residing in the occupied part of France. It is accompanied by a very large and probably quite complete bibliography of the subject and some of the bibliographic footnotes indicate a considerable amount of critical examination of the papers cited.

The article does not contain much of Belaiew's recent contributions to the subject which entered his book on the "Crystallization of Metals" (in French), but carries a considerable number of simple calculations on the sizes of the pearlite grains and lamellae of the carbide resulting from the tempering of both pearlite, medium and fine, bainite and martensite.

Steel metallurgists might be especially interested in the hardness versus dispersion diagram which begins with carbide particles averaging 4000 $\mu\mu$ in diameter and ends at the diameter of 0.67 $\mu\mu$ —at molecular dimensions. Up to the dispersion of 9.5 we are in the field of microscopic visibility, beyond that in the field of pure computation. The latter might be reasonably correct, but this abstracter expresses grave doubts in regard to the possibility of the existence of particles of molecular dimensions. A dispersion number of 18 is more likely to form a physical limit than Belaiew's dispersion number of molecular dimensions. Accurate computations in the field of grain size will require data on the dimensions of the unstrained unit crystallographic cell of far greater accuracy than at present available.

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December, 1948; Page 883

Low-Temperature Properties of Solders

SOLDERS are now being used extensively in equipment for refrigeration and liquefaction. However, until recently, nothing had been published concerning the properties of solders at subzero temperature. The two papers abstracted here were the only sources of information about the low-temperature tensile properties of solder until publication of the article on p. 843 in this issue.

Sn-Pb and Ag-Cu-Zn*

THE specimens were machined from ingots made from 99.98% lead, 99.74% tin, 99.9% copper and 99.41% silver. Tensile tests were made at +17, -196 and -253° C. (+63, -321 and -424° F.). The test results are summarized in Table I.

*Abstracted from "Mechanical Properties of Metals and Alloys in Tension at Low Temperatures (-196 and -253° C.). Part IV—Solders", by V. I. Kostenetz and A. M. Ivanchenko, *Journal of Technical Physics* (U.S.S.R.), Vol. 16, No. 5, 1946, p. 551-554 (in Russian).

The tensile strength of all the solders increased as the temperature decreased.† In general, the ductility decreased with decreasing temperature. The soft solder containing 25% tin and 75% lead showed the best combination of strength and plasticity at -321 and -424° F. The addition of copper and antimony had only slight effect on the low-temperature properties of the soft solders. Of the two silver

†EDITOR'S NOTE: However, these investigators missed entirely the maximum points in the curves of strength versus temperature for high-tin alloys, because no data were obtained between +63 and -321° F. For example, compare Table I with the curves on p. 886, giving data obtained by H. S. Kalish and F. J. Dunkerley for three tin-lead alloys containing 1.0, 5.0 and 50% Pb.

brazing alloys tested, the one containing 45% silver had the better combination of properties.

Tin-Rich Tin-Lead‡

THE specimens were made from tin and lead of extremely high purity. Pure tin and tin-lead alloys, containing 0.01, 0.05, 0.1, 0.35, 1.0, 1.66, 5.0, 10.0, 38.7 and 50.0% lead, were tested at eleven temperatures between +27 and -196° C. (+80 and -321° F.). All castings were homogenized at 170° C. for 200 hr.


(Continued on p. 886)

‡Abstracted from "The Low-Temperature Properties of Tin and Tin-Lead Alloys", by H. S. Kalish and F. J. Dunkerley, *Metals Technology*, Vol. 15, No. 6, September 1948.

Table I—Low-Temperature Tensile Test Data on Solders

MATERIAL	TENSILE STRENGTH, PSI.			REDUCTION OF AREA, %		
	+63° F.	-321° F.	-424° F.	+63° F.	-321° F.	-424° F.
90 Sn, 10 Pb	7,700	16,000	20,000	*	18	2
60 Sn, 40 Pb	8,000	17,000	21,000	49	6	1
50 Sn, 50 Pb	7,700	20,000	26,000	66	3	6
33 Sn, 67 Pb	7,700	20,000	26,000	76	19	11
25 Sn, 75 Pb	7,400	18,000	24,000	87	27	21
45 Ag, 30 Cu, 25 Zn	61,000	74,000	75,000	33	23	15
70 Ag, 20 Cu, 10 Zn	34,000	54,000	57,000	15	21†	13

*Wedge-shaped fracture. †The data showed a large spread.



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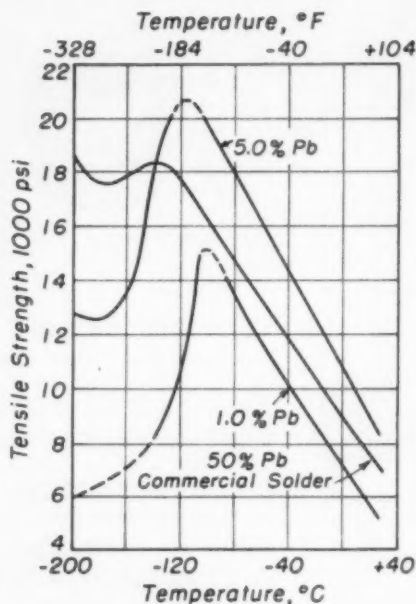
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Tin-Rich Solders

(Continued from p. 884)

before being tested. The results of tests were reported in 15 graphs.

Pure tin and all the tin-lead alloys containing up to 5.0% lead become brittle at about -165° F., except for the 1.0 and 1.66% lead alloys, which give brittle fractures at higher temperatures. Alloys with more than 10% lead remain ductile at successively lower temperatures as the lead content increases.



Tin and tin-lead alloys increase in strength as the temperature is decreased before embrittlement, as shown for three alloys in the accompanying diagram.

A 50-50 tin-lead commercial solder was brittle below -240° F., but at this temperature the tensile strength increased to 18,500 psi. Solders of higher lead content will become brittle only at successively lower temperatures until, at about 70 to 80% lead, the tin-lead alloys should remain ductile down to temperatures near absolute zero.

New Welding Methods Using Inert Gas

IN THE conventional process known as inert-gas-shielded arc welding, the heat of welding is produced by an arc between a single, nonconsumable tungsten electrode and the work. Inert gas (argon or helium) is fed through a nozzle surrounding the electrode in the head of the torch, and flows out

(Continued on p. 888)

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Shielded Welding

(Continued from p. 886)

to blanket the electrode, the arc and the weld puddle, in order to exclude the oxidizing atmosphere.

Two new welding processes using inert gas for shielding were described at the annual meeting of the American Welding Society in Philadelphia, Oct. 25 to 29, 1948. These new processes, known as inert-gas-shielded metal-arc welding and inert-gas-shielded-arc spot welding, are discussed briefly in the two abstracts that follow.

Inert-Gas-Shielded Metal-Arc Welding

J. S. Sohn and A. N. Kugler (Air Reduction Co.) reported the development of the inert-gas-shielded metal-arc welding process, in which a consumable filler wire is deposited continuously in a controlled gas atmosphere. The filler metal carries welding current, and an arc is maintained between the end of the wire and the work. Power may be supplied from a standard welding generator. The equipment developed includes a manual gun and an automatic welding head.

This process is particularly adapted to the welding of aluminum and aluminum alloys. Direct current, reversed polarity, gives the necessary arc conditions. Welding-grade argon is a satisfactory shielding gas.

One of the principal features of this process is the use of very high current densities. Current densities twelve times those used in ordinary metal-arc welding have been successfully employed. With such high current densities, high rates of filler metal feed must be provided in order to make satisfactory welds. Filler wire speeds for the manual welding gun range from 100 to 300 in. per min.

For the manual welding of aluminum as currently provided in this process, only three sizes of filler metal are needed — $\frac{3}{16}$, $\frac{1}{8}$, and $\frac{3}{32}$ in. With these filler metal sizes the over-all current range is 70 to 450 amperes, which permits welding of metal thicknesses from $\frac{1}{8}$ to 2 in., or more.

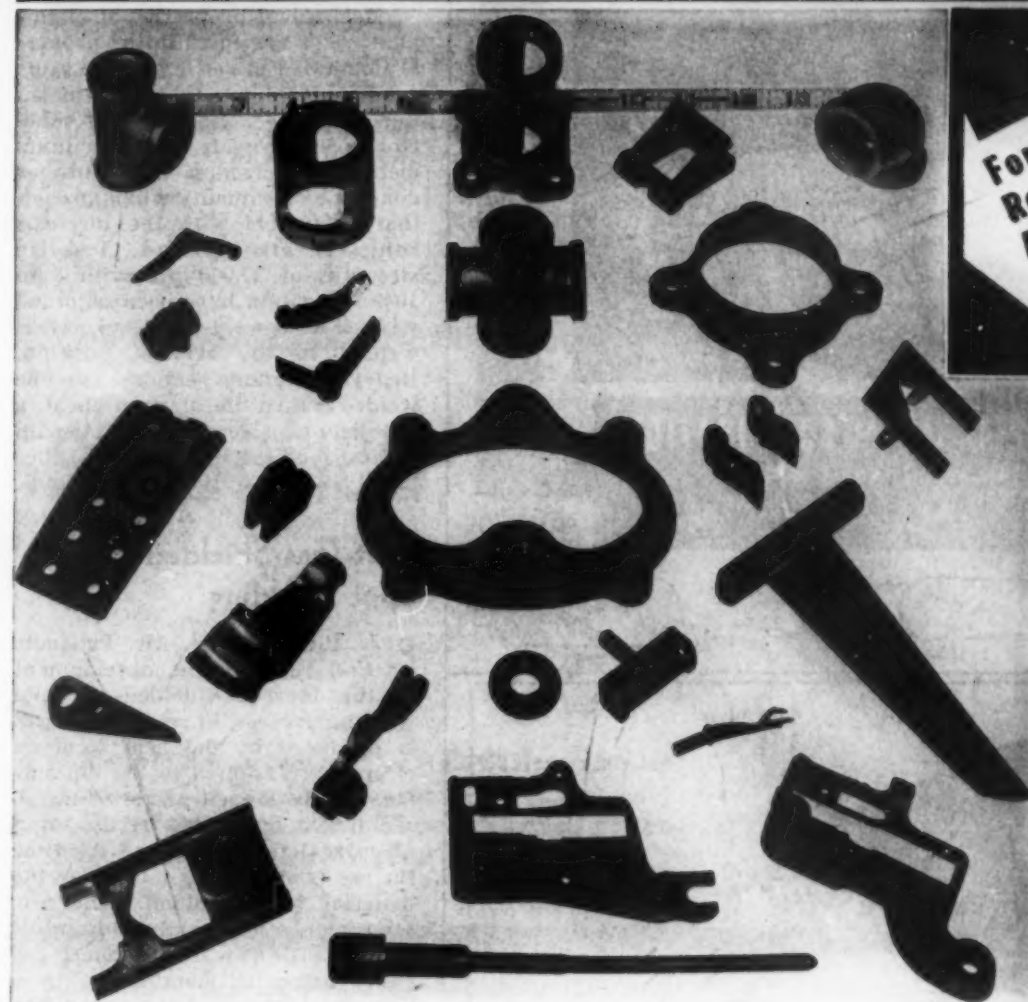
The selection of filler metal composition is governed by the metal to be welded. For the welding of 2S and 3S alloy, 2S wire is recommended; for 52S and 61S,

(Continued on p. 890)

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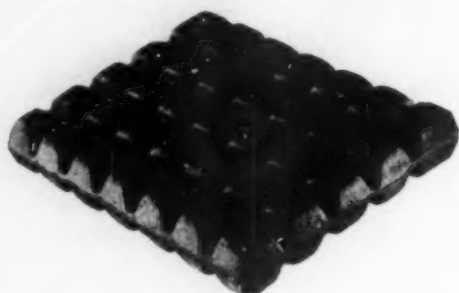
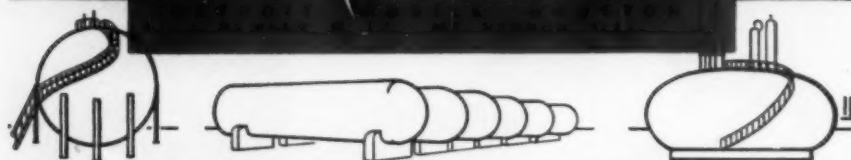
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Shielded Welding

(Continued from p. 888)

either 43S or 716 may be used. This process should reduce some of the limitations now encountered in the welding of heat treatable aluminum alloys.

Mechanical properties of aluminum welds are equal to the properties obtained in castings of the same composition as the filler metal. Generally the welds will be superior to welds made with aluminum metal-arc electrodes. Under proper conditions the quality will approach that obtainable with the inert-gas tungsten-arc method. Tensile strengths of 27,000 psi. with 8 to 10% elongation have been obtained with 43S all-weld-metal specimens welded in the vertical position. Heavy aluminum sections may be welded with little or no preheat, a practice in sharp contrast with other processes in which the filler metal is added by indirect methods.

Inert-Gas-Shielded-Arc Spot Welding

F. J. Pilia (Linde Air Products Co.) reported the development of the inert-gas-shielded-arc spot welding process, in which the weld is produced by the heat from an electric arc applied to the top surface of two lapped pieces of metal. Welding is controlled by the input of current to the arc and the time the arc is allowed to remain on the material being welded. The arc, tungsten electrode and weld puddle are shielded with an inert gas [only argon is mentioned], in a manner similar to that employed in conventional inert-gas-shielded arc welding.

In resistance spot welding, the weld is produced by the effect of high current and pressure applied to the electrodes. In inert-gas-shielded-arc spot welding, the weld is produced with fusion alone. Because no forging pressure is required on the fused nugget, no backup or access to the reverse side of the weld is necessary. This accounts for one of the major advantages of the gas-shielded spot welding process—the ability to produce spot welds from one side of the joint only.

The apparatus required for gas-shielded spot welding consists of a standard 60-cycle a-c. metal-arc welding transformer with a built-in

(Continued on p. 892)

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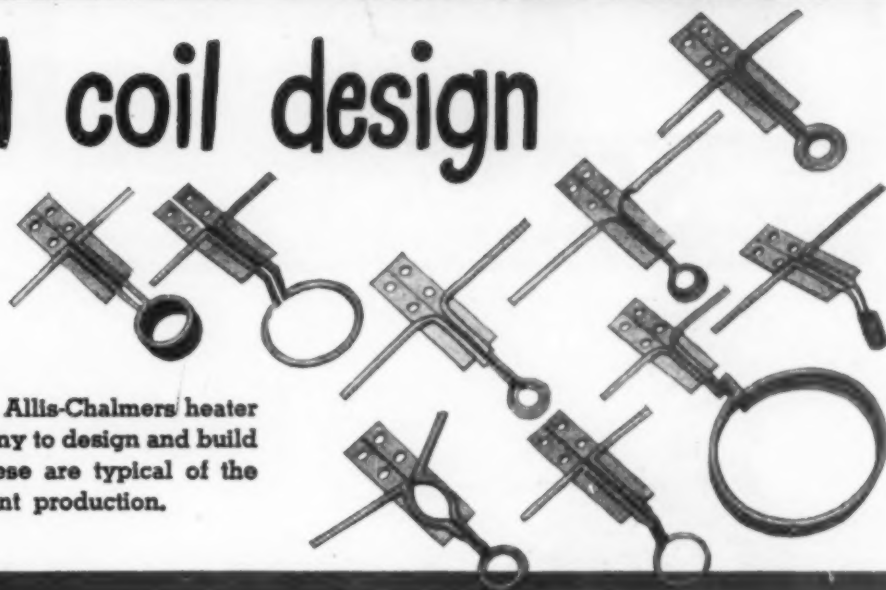
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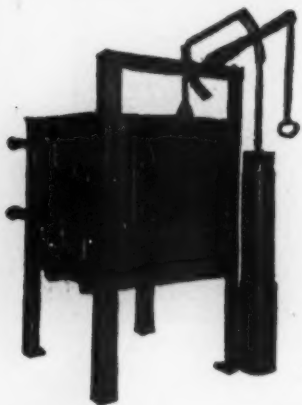


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Shielded Welding

(Continued from p. 890)

high-frequency unit. A timing device similar to that used for resistance spot welding is employed, except that longer time cycles are required for the arc, and instead of circuits for "squeeze" and "hold", there is a circuit for argon shutoff delay.

The tool used for making the weld consists of a gun of molded plastic about the size of a Colt "45" automatic, and weighing 2 lb. The gun is equipped with a water-cooled copper cup, which bears against the work. A tungsten electrode from which the electric arc issues is located centrally in this cup. It is spaced back from the end of the cup by about $\frac{1}{8}$ in. This setting varies with the material being welded and is adjustable.

The application of the process to the assembly of automobile bodies is discussed in detail in the paper. The auto body is designed primarily around the resistance spot welding process with some use of oxy-acetylene welding, metal-arc welding and flash welding. Metal-arc welding in body construction is used where resistance spot welding is not feasible because of the size of the machine required, the obstruction of the work, or inability to reach the opposite side of the work for resistance welding backup. Gas-shielded spot welding is satisfactory for most of the auto body welding that has been done by the metal-arc process.

In addition to the automotive operations, the process has been used successfully in attaching light sections to heavy sections. For instance, a 0.040-in. sheet can be attached to a 4-in. thick section in the various metals. Other applications are mentioned in the paper.

Hydrogen in Aluminum*

THE solubility of hydrogen in liquid aluminum of high purity has been redetermined by a standard method. The solubility in solid aluminum, usually reported as nil, has been determined with reasonable accuracy.

(Continued on p. 894)

*Abstracted from "The Solubility of Hydrogen in Liquid and Solid Aluminum", by C. E. Ransley and H. Neufeld, *Journal of the Institute of Metals*, Vol. 74, 1948, p. 599 to 620.

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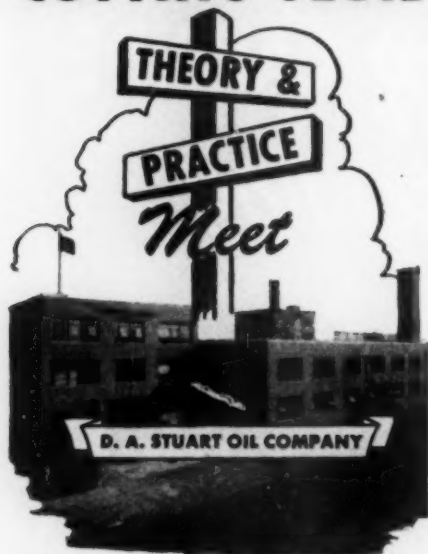
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Gas in Aluminum

(Continued from p. 892)

The solubility values at 1 atm. pressure are as follows:

TEMPERATURE, °C.	SOLUBILITY, CU. CM. PER 100 G.
0 (estimated)	1×10^{-7}
300	0.001
500	0.012
660 (solid)	0.036
660 (liquid)	0.6
700	0.9
750	1.2
850	2.1

The values for solubility in liquid aluminum are much higher than almost all previous determinations.

It is shown that the porosity of a sand-cast test bar of commercial-purity aluminum is a linear function of the hydrogen content.

The volume of hydrogen dissolved in commercial melts of aluminum and the common aluminum alloys is about 0.1 to 0.6 cu.cm. per 100 g. A melt of aluminum held at 725° C. and containing 0.3 to 0.4 cu.cm. per 100 g. of hydrogen will have an "internal pressure" of about 0.1 atm.; if the gas content is reduced to 0.1 cu.cm. per 100 g., the internal pressure will decrease to 0.01 atm. This means that if the metal is degassed by passing a stream of bubbles of neutral gas through it, the theoretical maximum rate of removal of hydrogen will initially be 10% of the volume of gas passed, but this will decrease to 1% as the hydrogen content approaches 0.1 cu.cm. per 100 g. Because of the limited time of contact of the bubbles with the melt, and the operation of other factors tending to restrict the diffusion of hydrogen into them, it is very unlikely that the percentage of hydrogen carried out of the metal even approaches the theoretical value. The practical difficulty in reducing the hydrogen content to a low level by flushing with a neutral gas is thus readily understood.

Stress Relief at Low Temperature*

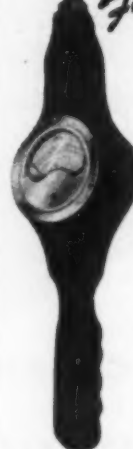
THE most efficient but rarely feasible way of relieving stresses in welded structures consists of heating the whole weldment to a temperature where plasticity is sufficient to eliminate the whole complex of stresses. (To p. 896)

*Abstracted from "Low-Temperature Stress Relieving", by T. W. Greene. Presented before the International Acetylene Assoc., May 3, 1948.

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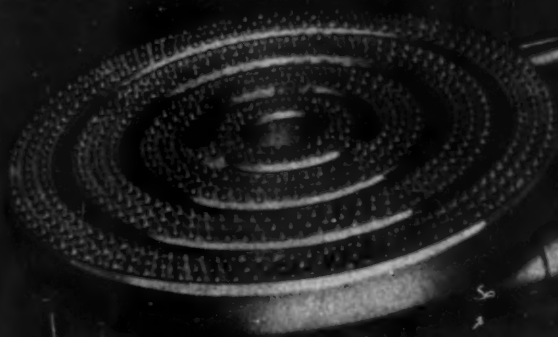
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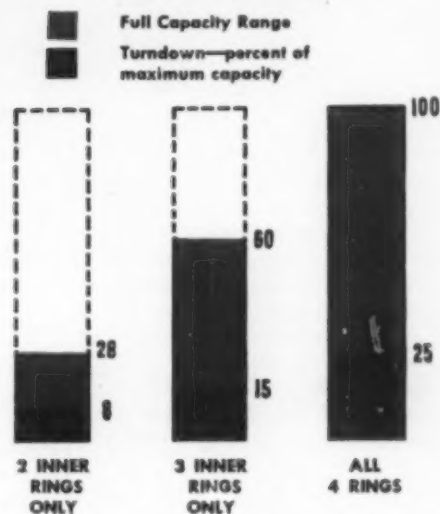


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Stress Relief

(Continued from p. 894)

Stretching the structure so the metal adjacent to the weld is elongated to just below its yield point could also produce a plastic stretch in the weld, and this would eliminate the contraction resulting from the greater cooling of the weld in comparison to the metal welded. This, however, is even less feasible than general heat treatment.

The author describes a method which consists of a controlled heating of the metal outside the weld, thus stretching it thermally and the weld mechanically. When the thermal elongation disappears (on cooling) the mechanical plastic elongation of the weld remains; in that manner the residual stresses are eliminated.

The method consists in heating up a certain width of the metal equally on both sides of (and at a definite distance from) the weld. The metal is allowed to reach a temperature of 350° F. and immediately cooled by jets of water following the heating flames at a given constant distance. The weld is kept cool continuously by a jet of water.

The method was worked out from a thorough study of the stresses of welding, and the stresses remaining after relief at low temperatures. For this purpose a welded plate 7x4 ft. prepared of 7x2-ft. halves was fitted with electric resistance gages in its central part, and strips carrying the gages cut out (trepanned) by drilling and sawing, so as to separate them from the body of the plate and so eliminate the stresses between the strip and the plate. The change in stress indicated the previously existing average stress in that strip. The same procedure applied to the plate relieved at low temperature yielded the magnitudes of stresses still retained in such strips. It was found that the remaining stresses had been reduced to a nearly negligible quantity.

Tables are presented in the text giving the distances of the heating jets from the weld and the velocities of their travel for plate thicknesses between ½ and 1¼ in., and optimum conditions of operation.

The paper also describes the actual applications of the method to the construction of tankers, pressure vessels and to the water tank of a 10-million cu.ft. gas holder — the largest in the world.



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Gaging Thin Films*

THE problem of gaging the thickness of extremely thin films has long been a headache in industry.

For example, the average thickness of a sheet of Pliofilm, a Good-year product, is about 0.001 in. (one mil). Mechanical gages are difficult to use since the tiniest change in the pressure of the gage on the film will give a false reading. Present gages read with only dubious accuracy to 0.0001 in.

The usual mechanical gage compresses the film between a metallic foot and an anvil. The theory is that the weight of the foot will provide the same pressure for each test, but actually the softness of the film will affect the result.

Various refinements may be incorporated, but one difficulty about using all such gages in conjunction with a production line is that either the progress of the film must be halted for gaging purposes or a piece of film must be cut from the sheet and subjected to gaging.

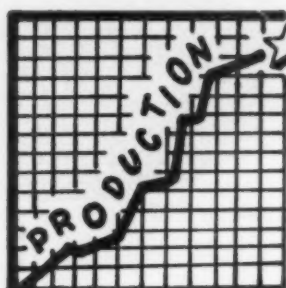
*A note concerning research by Goodyear Tire and Rubber Co., Akron, Ohio, contained in the Fourth Semiannual Report of the United States Atomic Energy Commission.

A new radioactive gage, using radiocarbon (C^{14}), makes it possible to gage a sheet of Pliofilm or other film continuously as it comes from the rolls. No mechanical contact with the film is required.

The film passes through a slot in the gage. Below this slot there is a small bit of C^{14} . Above the slot there is an ionization chamber in which is produced a minute electrical current by the rays from this radioactive carbon. This current is amplified by subminiature vacuum tubes to the point where it is sufficiently large to operate an ordinary electric meter which will then indicate the strength of the rays. Further amplification could be used to actually control the film gage by regulation of the speed of the production machinery.

Radioactive carbon C^{14} gives off only a weak stream of beta rays. These are electrons like the electrons released by the filament of an ordinary radio vacuum tube. The number of electrons getting through the film depends on its thickness, and the meter can be graduated to read directly in thickness of film.

The new gage reads to a hundredth of a mil (0.00001 in.), but it is hoped to attain an accuracy ten times greater.



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	TYPE	Max. Temp.	Series	Recommended for	Bulletin No.
TILTING	CRUCIBLE	2500°F.	TCR	Melting Brass and Copper	TC-45
	POT	1400°F.	TCA	Melting Aluminum	TC-45
		1400°F.	TP	Melting Aluminum and Magnesium	TP-45
STATIONARY	CRUCIBLE	2500°F.	CR	Melting Brass and Copper	CR-45
		1400°F.	AC	Melting Aluminum	AC-45
	POT	1400°F.	AM	Melting Aluminum	AM-45
		1200°F.	SM	Melting Soft Metals	AM-45
		1000°F.	SA SAV	Melting Soft Metals	SV-45

